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AFWAL-TR-80-4175

RESEARCH TO CONDUCT AN EXPLORATORY EXPERIMENTAL AND ANALYTICAL INVESTIGATION OF ALLOYS

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United Technologies Corporation Pratt & Whitney Aircraft Group
Government Products Division
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November 1980

Final Report for Period April 1979 to August 1980

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This technical report has been reviewed and is approved for publication.

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Alloys based on the TiAl (Gamma) and Ti_3Al (alpha-two) intermetallic phases have attractive elevated temperature properties for gas turbine applications. However, certain processing and property problems will have to be overcome before such materials are qualified for the projected applications.

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The current program evaluated processing and heat treatments for three alpha-two and three gamma alloys. The basic compositions were Ti-Al-Nb (V) and Ti-Al-V, respectively. The effect of various interstitial elements, H_2, O_2 and C were also evaluated. Mechanical properties measured were smooth and notched tensile strength up to 650C (1200F), room temperature notched stress rupture, and creep-rupture over the range 593C (1100F) to 982C (1800F).

Test results indicate that useful engineering alloys are evolving in both systems, although melting and processing of the gamma phase alloys is not as advanced as that of alpha-two alloys.

FOREWORD

This report covers work performed on Contract F33615-75-C-1167 (P00010), Task No. 7021-01-68, for the period April 1979 to August 1980.

The investigation was conducted by the Commercial Products Division Pratt & Whitney Aircraft, East Hartford, Connecticut under the technical direction of Dr. H. A. Lipsitt, AFWAL/MLLM, Wright-Patterson Air Force Base, Ohio.

Dr. M. J. Blackburn was program manager and Mr. M. P. Smith was the responsible engineer. The experimental assistance of Mr. D. R. Haase is gratefully acknowledged. Dr. J. C. Williams, Carnegie-Mellon University, coordinated the electron microscope studies on the gamma alloys.

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SECTION I

INTRODUCTION

Major advances in gas turbine engine technology have been based upon the development of improved nickel and cobalt base superalloys and the conventional alpha-beta titanium alloys. The high temperature properties of current titanium alloys have improved to the point where the majority of compressors in advanced engines utilize these lightweight materials. Further extension of the use of lightweight materials to higher temperature structures in the turbine or afterburner sections of the engine is highly desirable, but as conventional titanium alloys have only limited scope for further improvement, few new applications can be anticipated. Extending the use of titanium base alloys may be possible if a new approach is adopted. Investigations performed in the mid-1950's by McAndrew and Kessler[1] at Armour Research Foundation identified several unique characteristics associated with the TiAl (gamma phase) of the titanium-aluminum system. In particular, the intermetallic alloys were found to possess high specific strengths at elevated temperatures, good oxidation resistance and a high specific modulus of elasticity. Unfortunately, other characteristics such as essentially zero room temperature ductility, low impact strength and poor formability led to the abandonment of this research. Similar studies of alloys containing the Ti_3Al (alpha-two) phase were conducted by McAndrew and Simcoe in the early 1960's[2,3]. Again, attractive elevated temperature properties were found along with limited room-temperature ductility.

Design analysis and payoff studies conducted on these materials have indicated significant weight savings would be possible in a wide range of engine applications if adequate engineering properties were developed. Turbine rotor weight savings from 30 to 40 percent (three to five percent of engine weight) would be achieved with widespread application of the titanium aluminides in rotating hardware; weight savings of up to 16% could be achieved in engine applications to static structures such as vanes, cases and bearing supports. Beyond the immediate and obvious savings in engine weight, it is possible to translate these benefits into significant fuel savings with attendant effects on operating cost.

With the advent of more advanced technologies in physical and process metallurgy by the early 1970's, it was decided to reinitiate studies of alloys based on these intermetallic compounds. During the past six years, Pratt & Whitney Aircraft has participated in AFWAL/Materials Laboratory sponsored programs directed

toward the understanding and exploitation of alloys based on both the Ti_3Al and TiAl phases. Contract F33615-C-75-1167 has performed alloy and process development on Ti_3Al alloys; Contracts F33615-C-74-1140 and F33615-C-75-1166 have conducted similar development efforts on TiAl alloys. The goals in all programs were to identify alloys from the base systems that exhibited useful properties and develop processing methods. Results from the first parts of these studies have been summarized in interim reports issued during 1978[4,5] and 1979[6].

SECTION II

PROGRAM APPROACH

1. Introduction

The overall objective of the program was to screen a restricted range of alloys (TiAl and Ti_3Al base) and processing sequences with the best chance of yielding useful engineering materials. The technical effort was divided into four tasks as follows:

Task I	Alloy Definition/Trace Element Effects
Task II	Alloy Preparation and Processing
Task III	Heat Treatment Studies
Task IV	Property Determinations

A summary of the approach for each task is presented in the following sections.

2. Task I - Alloy Definition/Trace Elements

a. Ti_3Al System

Previous Ti_3Al alloy development efforts had shown that Ti-Al-Nb alloys had the most promise for good tensile and creep rupture properties[7]. All indications were that a niobium content of 12-15 atomic percent provided the best property balance, as shown in Figure 1, and that some of the niobium in these alloys could be replaced by vanadium. Advantages of such substitution include lower alloy density, more readily available master alloys, and lower cost. The current program concentrated on defining the properties of alloys with compositions included in the shaded area of Figure 1. Additional exploration of the vanadium substitution effect was also undertaken.

One of the inconsistencies in the previous titanium aluminide investigations has been the poorer properties, specifically room temperature ductility, that has been found in scaled-up alloys when compared with the smaller laboratory size heats. It has been suggested that this difference may be traced to differences in oxygen content between the materials. Therefore, in this program, a more systematic study was conducted to quantify any oxygen effect and provide guidance for specification

*Unless specified, all compositions are given in atomic percent

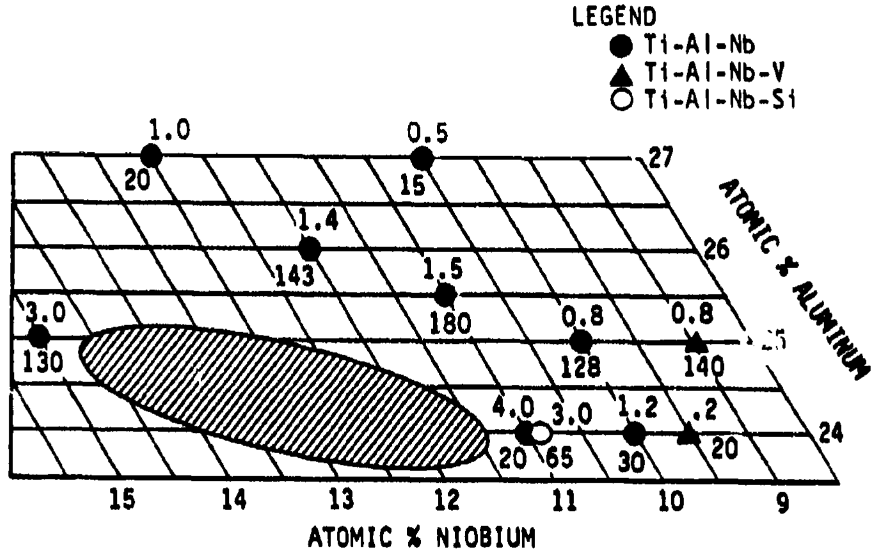


Figure 1

Room temperature tensile ductility (top no.) and 650C/379 MPa (1200F/55 ksi) creep rupture life (bottom no.) of forged plus solution treated Ti-Al-Nb and Ti-Al-Nb-X alloys as a function of their alloy content. The shaded area indicates the region evaluated in the current study.

limits. For one selected base alloy, two additional ingots with different oxygen levels were prepared to assess the influence of this element on processability and properties. Hydrogen effects have also not been studied in Ti_3Al type alloys previously and, therefore, were included in the present investigation.

b. TiAl System

Prior studies have shown that alloys based on the two phase $\alpha_2 + \gamma$ system (46-50% aluminum) offer potential as useful engineering material. The best combination of room temperature tensile strength and ductility coupled with elevated temperature creep rupture properties were found for an aluminum content of 48%[4]. Various alloy additions to this base were studied, but the only useful ternary addition found was vanadium. Additions of 1-2% were shown to improve tensile ductility, especially at intermediate temperatures[4]. Vanadium did not appear to influence stress-rupture properties to any large extent, but small additions of carbon were shown to produce an increase in rupture capability by a factor of up to six.

Consequently, the current program concentrated on the evaluation of vanadium containing alloys, one of which was further modified with a carbon addition. As with Ti_3Al base alloys, very little is known about the effects of the interstitial elements oxygen and hydrogen on the properties of TiAl base alloys. Thus, a similar study to that undertaken for Ti_3Al alloys was included for this class of alloys in this part of the program.

3. Task II - Alloy Preparation and Processing

The melting and processing of Ti_3Al base alloys has developed to the extent that conventional titanium alloy practice can be used. Conventional techniques were employed to reduce the ingot material to forged pancakes. Methods for producing large sections of TiAl are more restricted but, in the past year, TMCA developed effective procedures for large ingot melting. Production of ingots, using these methods, followed by isothermal forging was selected as the processing method for TiAl type alloys.

4. Task III - Heat Treatment Studies

a. Ti_3Al System

Heat treatment of Ti_3Al base alloys can be used to produce quite large variations in properties. The best ductility at low temperatures and the best creep rupture capability is produced by a beta solution treatment, but the cooling rate from the beta phase field is also of crucial importance. By controlling the cooling rate to give a fine Widmanstätten structure, the best property balance is achieved. The use of quenching (and holding) at intermediate temperatures has shown some promise of giving both the desired structure and improved properties. Aging in the temperature range 700-900C (1300-1600F) also had been found to increase ductility with some reduction in stress-rupture capability. As these latter two effects have been shown on large section sizes, they have important practical implications and were further studied in this program.

b. TiAl System

All of the alloys studied in this program lie in the two phase gamma/alpha-two phase field. However, the phase transformations in this type of alloy are not clearly understood and, therefore, a limited study to clarify the structures formed was undertaken as part of this program.

In the past, it has been considered that when starting with a forged product with an exquiaxed structure, the only major variables were grain growth during solution treatment with minor changes produced on subsequent aging. However, it has been realized more recently that the acicular structures often observed in castings and extrusions are formed by exposure to temperatures about 1342C (2450F), probably in a beta phase field. These acicular structures offer improved creep capability in some alloys. Thus, a somewhat wider range of solution treatments were included in the present program.

5. Task IV - Property Determinations

The main test methods have remained constant through all the programs and include hardness, bend, tensile, creep-rupture, and low cycle fatigue testing. However, in order to evaluate the influence of interstitial elements, it was necessary to add notch testing to the program. Notch tensile behavior was evaluated for oxygen containing alloys to monitor any major change in brittleness characteristics. Hydrogen effects are usually time dependent and are also more readily detected in notched specimens. Therefore, room temperature notch rupture tests, uploaded after pre-determined time intervals until rupture occurs, were employed to study the influence of this element.

SECTION III

EXPERIMENTAL DETAILS

1. Task I - Alloy Selection

Alloy chemistries were chosen to cover the ranges of interest previously defined in Section 2.2. Aim compositions for the ten alpha-two and gamma alloys are listed in Table 1.

2. Task II - Alloy Preparation and Processing

a. Melting - Eight of the ten required ingots were double consumably vacuum arc melted at the Henderson Technical Laboratory of TIMET, Henderson, Nevada. The alpha-two ingots were approximately 10 cm (4 inches) in diameter by 18 cm (7 inches) high and weighed approximately 7-8 Kg (15-18 lbs) (Figure 2). The gamma alloy ingots were approximately 15 cm (6 inches) in diameter by 40 cm (16 inches) high and weighed about 25 Kg (55 lbs) (Figure 3).

The low oxygen content ingots were prepared in Pratt & Whitney Aircraft laboratories using high purity electrolytic titanium sponge. Eight 2.5 cm (1 inch) diameter by 10 cm (4 inch) long drop castings were prepared for each alloy. Four castings were welded together using AMS 4901 wire to form two primary electrodes per alloy. These were then consumably melted in a Leybold-Hereaus unit to produce a final ingot about 6 cm (2.5 inch) in diameter by 15 cm (6 inch) long, weighing approximately 1 kg (2.2 lbs). A typical primary and final laboratory ingot is shown in Figures 4 and 5, respectively.

b. Ingot Evaluation - TIMET Material

Visual examination of the eight TIMET ingots revealed no surface connected cracks.

Radiographic examination revealed the presence of a small shrinkage cavity (pipe) about 1-2 cm (0.5-0.75 inches) round near the top surfaces of the alpha-two ingots. The gamma alloy ingots also exhibited this type of pipe shrinkage but, in addition, the Ti-48Al-1V (V-5767, V-5768) ingots exhibited

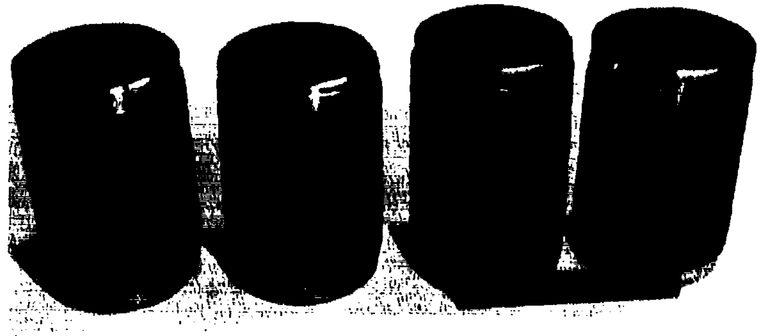


Figure 2

As-received Ti_3Al (alpha-two) ingots produced by Titanium Metals Corporation of America.

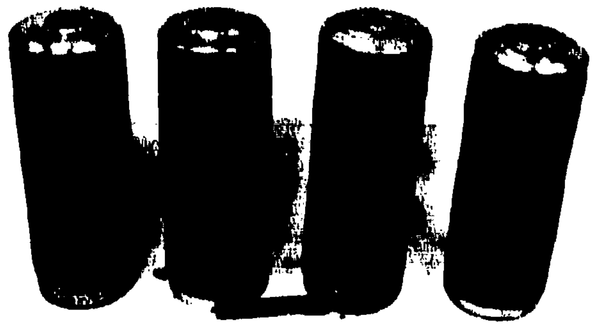


Figure 3

As-received TiAl (gamma phase) ingots produced by Titanium Metals Corporation of America.

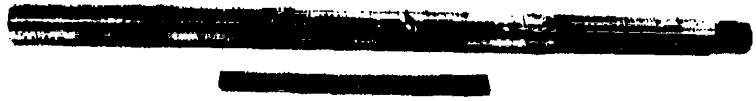


Figure 4

Primary electrode for MERL ingot consisting of four drop castings welded together.

Figure 5

Final MERL low interstitial content ingot consisting of two primary electrodes melted together.

Table 1

Alloy Aim Compositions Selected For Contract F33165-75-C-1167 in Atomic % (Weight %)

Type	Al	Nb	V	O_2	Other[1]
Alpha Two	25.0(14.0)	14.0(26.6)	-	1000 ppm[2] max	-
	25.0(14.0)	14.0(26.6)	-	1500 ppm	-
	25.0(14.0)	14.0(26.6)	-	$\leq$500 ppm	-
	25.0(14.0)	10.0(19.6)	4.0(4.3)	$\bar{1}$000 ppm max	-
	24.5(13.6)	13.0(24.8)	-	1000 ppm max	-
Gamma	48.0(34.5)	-	1.0(1.3)	1000 ppm[2] max	
	48.0(34.5)	-	1.0(1.3)	1500 ppm	
	48.0(34.5)	-	1.0(1.3)	$\leq$500 ppm	
	48.0(34.5)	-	1.0(1.3)	$\bar{1}$000 ppm max	0.1(0.05)C
	49.5(35.8)	-	0.5(0.7)	1000 ppm max	-

(1) Balance Titanium

(2) Hydrogen added to specimens of this composition by thermal techniques.

the presence of large "mottled" appearing areas. A section through the mottled area of V-5767 revealed that a considerable amount of shrinkage or gas porosity was present (Figure 6). The porosity was clean and did not appear to be surface connected. The gamma alloy ingots V-5769 and V-5770 each exhibited cracking in the lower third of the ingot, although it was difficult to determine if these were surface connected.

Chemical analysis of the alpha-two ingots revealed that the aim compositions were met with the exception of the high oxygen content baseline chemistry ingot (V-5764). It was found to contain only 0.083 to 0.103% oxygen rather than the goal 0.15% (Table 2). Since the number of specimens required for the interstitial study was small, it was decided to attempt to replace the V-5764 ingot with a smaller Pratt & Whitney Aircraft melted ingot, similar to that melted for the low interstitial content alloy.

Composition of the gamma alloy ingots is given in Table 3. All aim compositions were met with the following two exceptions; the high oxygen content ingot (V-5768) contained only 0.13% oxygen rather than the 0.15% aim, while V-5770 contained 48% aluminum rather than the 49.5% aim.

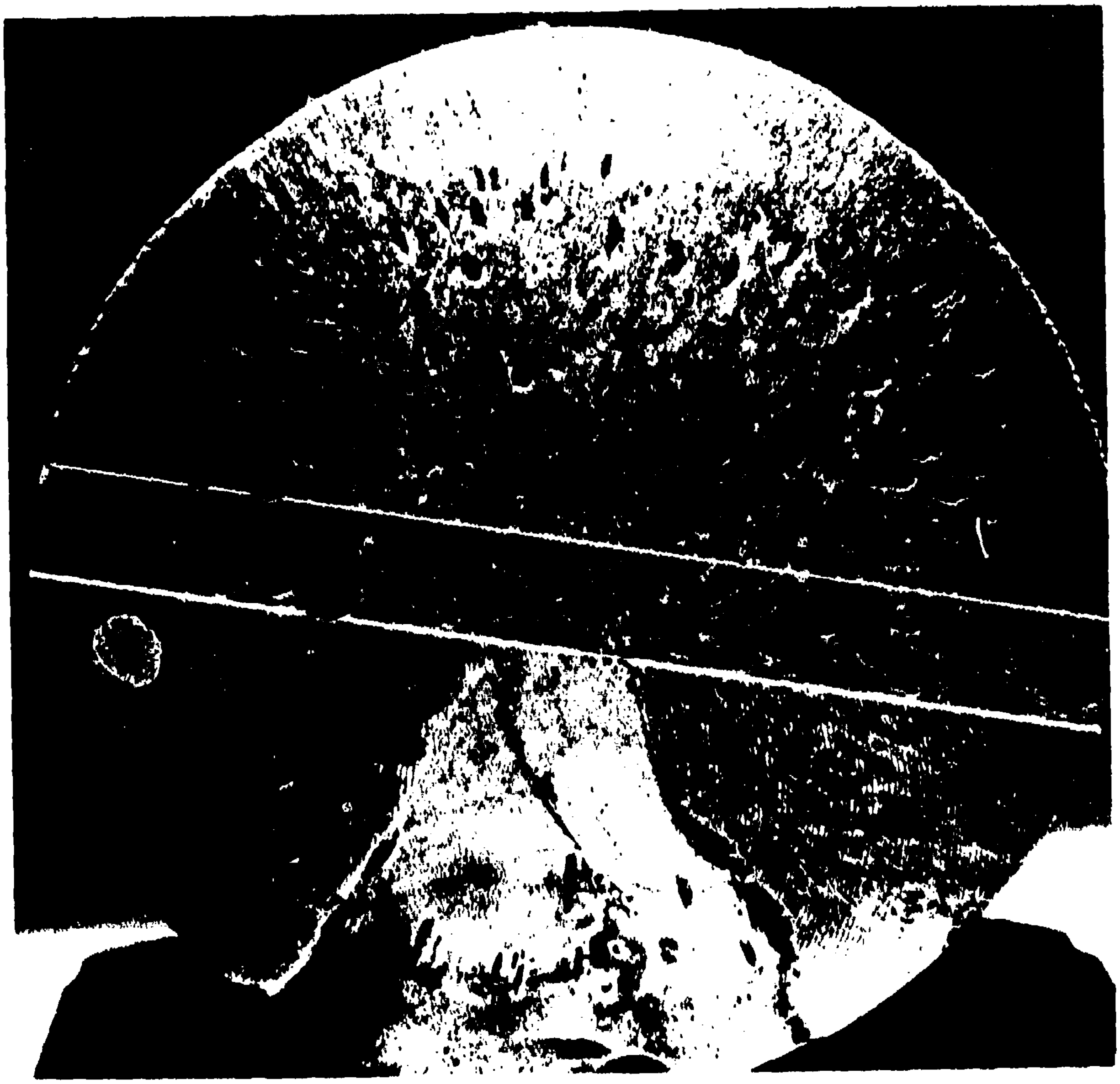


Figure 6 MAG: 1.2X

Cross section of macroetched Ti-48%Al-1%V ingot V-5767 showing porosity in the as-cast condition.

Table 2

Actual Compositions of TIMET Alpha-Two
Alloy Ingots for Contract F33615-75-C-1167
in Weight % (Atomic %) (1)

Heat	Sample(2) Type	Al	Nb	V	Fe	O_2	N_2
V-5763	Ingot	13.8	26.8	-	0.115	0.076	0.009
	Forging	13.6	26.7	-	0.157	0.076	0.010
	Aim	13.5-14.0 (25.0)	26.3-26.9 (14.0)	-	-	0.10 max	-
V-5764	Ingot	13.6	26.8	-	0.105	0.104	0.007
	Forging	14.1	26.5	-	0.13	0.083	0.009
	Aim	13.5-14.0 (25.0)	26.3-26.9 (14.0)	-	-	0.14-0.16	-
V-5766	Ingot	13.7	25.0	-	0.100	0.087	0.006
	Forging	13.6	25.3	-	0.125	0.076	0.008
	Aim	13.3-13.9 (24.5)	24.4-25.1 (13.0)	-	-	0.10 max	-
V-5810	Ingot	14.1	19.5	4.23	0.133	0.095	0.008
	Forging	-	-	-	-	-	-
	Aim	14.0-14.5 (25.0)	19.3-19.9 (10.0)	4.0-4.5 (4.0)	-	0.10 max	-

(1) Balance Titanium
(2) Ingot analysis conducted on turnings from top and bottom of ingot; forging analysis conducted on sections of forged pancake.

Table 3

Actual Composition of TIMET Gamma Alloy Ingots for Contract F33615-75-C-1167 in Weight % (Atomic %)(1)

Heat	Sample Type	Al	V	Fe	O_2	N_2	C
V-5767	Ingot	34.3	1.23	0.061	0.094	0.007	-
	Aim	34.0-34.5 (48.0)	1.0-1.5 (1.0)		0.10 max		
V-5768	Ingot	34.2	1.24	0.057	0.13	0.007	-
	Aim	34.0-34.5 (48.0)	1.0-1.5 (1.0)		0.14-0.16		
V-5769	Ingot	34.8	1.23	0.058	0.101	0.008	0.028
	Aim	34.0-34.5 (48.0)	1.0-1.5 (1.0)		0.10 max		0.02-0.04 (0.1)
V-5770	Ingot	34.6	0.62	0.060	0.099	0.006	-
	Aim	35.8-36.2 (49.5)	0.5-0.8 (0.5)		0.10 max		

(1) Balance Titanium

Ingot Evaluation - MERL Material

Visual and radiographic examination of the MERL melted ingots revealed no cracking, only a small amount of shrinkage porosity and pipe.

Chemical composition of the MERL melted ingots is given in Table 4. The attempt to melt the high oxygen content ingot was unsuccessful and both alpha-two ingots contained less than 0.05% (500 ppm) oxygen. Duo to an analytical error, it was originally thought that the high oxygen content ingot contained 0.17% O_2 but, based on properties obtained, was reanalyzed and found to be low in oxygen. For purposes of identification, the two low oxygen alpha-two ingots will be designated L and S.

c. Hot Isostatic Pressing (HIP)

The TIMET alpha-two ingots and low interstitial content MERL ingots were subjected to hot isostatic pressing (HIP) at 1180C/103 MPa/3 hours (2160F/15 ksi/3 hours) in the MERL laboratory unit. Subsequent re-examination by radiography revealed that the shrinkage cavities had sealed and there were no detectable subsurface voids. The four gamma alloy ingots were HIP'ed at 1204C/103 MPa/3 hours (2200F/15 ksi/3 hours) at Industrial Materials Technology, Inc., Woburn, MA. Subsequent radiographic examination revealed that the porosity in the Ti-48Al-1V ingots V-5767 and V-5768 had apparently been sealed. The cracks in ingot V-5770 were no longer visible; however, the carbon-containing ingot V-5769 still exhibited cracks. Fortunately, sufficient sound material was available to fabricate the required test specimens.

d. Forging

Four TIMET alpha-two ingots were conventionally beta forged into 3 cm (1.25 inch) thick x 30 cm (12 inch) diameter pancakes on open dies using the 1500 ton press at Wyman-Gordon Company, Millbury, MA.* The 10 cm (4 inch) diameter x 18 cm (7 inch) high ingots were upset 50% and redrawn at 1260C (2300F) twice prior to finish forging. The resulting pancakes are shown in Figure 7.

*Includes V-5764 which was not used in the program.

Table 4

Actual Compositions of the MERL Melted Ingots for Contract F33615-75-C-1167 in Weight % (Atomic %)(1)

Ingot(2)	Al	Nb	V	Fe	O_2(3)	N_2
Alpha-two Baseline "L"	13.7	26.6	-	0.023	.022	<0.002
Aim	14.0 (25.0)	27.0 (14.0)	-	-	0.15	-
Alpha-two	13.7	26.2	-	-	0.029	<0.002
Baseline "S"	14.0 (25.0)	27.0 (14.0)	-	-	0.05 max	-
Gamma Baseline	34.5	-	1.2	-	0.021	-
Low O_2	34.0 (48.0)	-	1.3 (1.0)		0.05 max	-

(1) Balance Titanium

(2) Alpha-two ingot "L" was an unsuccessful attempt to make a high oxygen content ingot.

(3) Average of four samples in all cases.

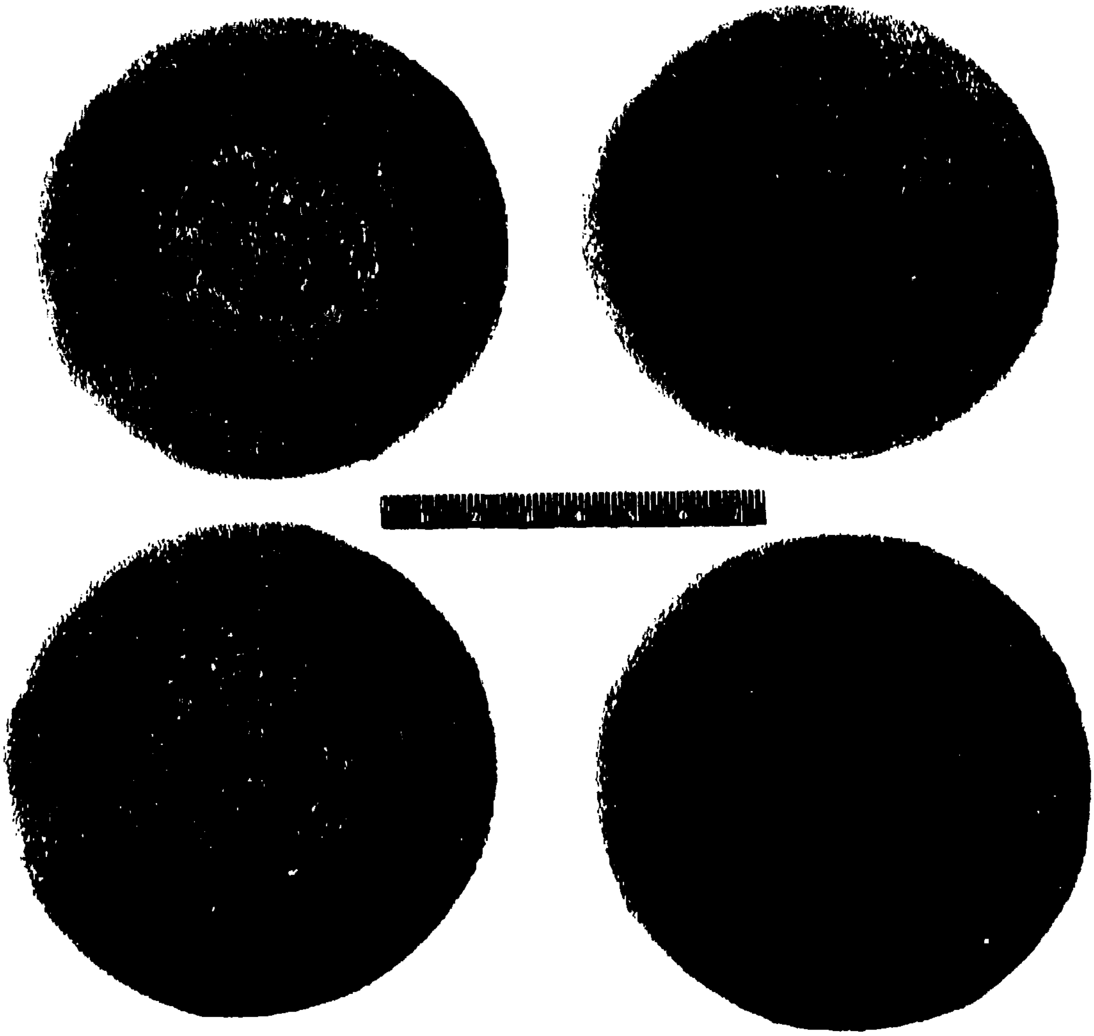


Figure 7 MAG: .25X

Alpha-two pancake forgings produced from ingots V-5763, V-5764, V-5810, and V-5766 by conventional forging at 1150C (2100F).

In the case of the gamma alloy ingots, cylindrical sections, about 13 cm in diameter x 13 cm high (5" x 5") were machined from areas which appeared to be sound based on the radiographs and lathe turned in preparation for forging. Some unhealed gas porosity was detected in V-5768 and a surface connected crack was detected in V-5769 after preparation. Since the cracking and porosity were not extensive, it was decided to attempt to forge the sections. They were isothermally forged on open dies using the 500 ton press at the Pratt & Whitney Aircraft Government Products Division (GPD) facility. The lathe turned sections were forged to a 50% reduction at 1150C (2100F). These upsets were sectioned and one of the halves finished forged to 2.5 cm (1 inch) thick at 1010C (1850F). The resulting pancakes are shown in Figures 8-11. Areas of porosity are visible on the cut faces of the first upsets, indicating that HIP was not totally successful in sealing porosity. Some shallow cracking was apparent on the cut faces of the final pancakes. They appeared to have been caused by opening up of residual cracks from abrasive cutting. The three MERL melted ingots were isothermally forged at the PWA-GPD facility. The gamma alloy ingot was forged into a pancake shape in the two step procedure outlined above. Some shear cracking occurred during the second upset at an interface where two of the small electrodes had been melted together. In order to avoid this problem, the alpha-two ingots were side upset at 1150C (2100F) in a one-step operation.

3. Task III - Heat Treatment Trials

Extensive heat treatment trials were conducted using small sections cut from the forged alpha-two and gamma alloy baseline composition pancake forgings. In the case of the alpha-two compositions, the following treatments were studied:

- A beta solution treatment followed by air cooling and aging
- A beta solution treatment followed by quenching into 815C (1500F) molten salt
- An alpha-two + beta solution treatment followed by oil quenching and aging

Solution temperatures were maintained at 14-28C (25-50F) above or below the transus as required. Transus determinations were conducted in a platinum tube furnace under an argon atmosphere. Accuracy of the furnace is $\pm$3C ($\pm$6F). All other heat treatments were conducted on oversize specimen blanks in an air atmosphere. The salt quenching medium was Swiftheat 1010 neutral salt.

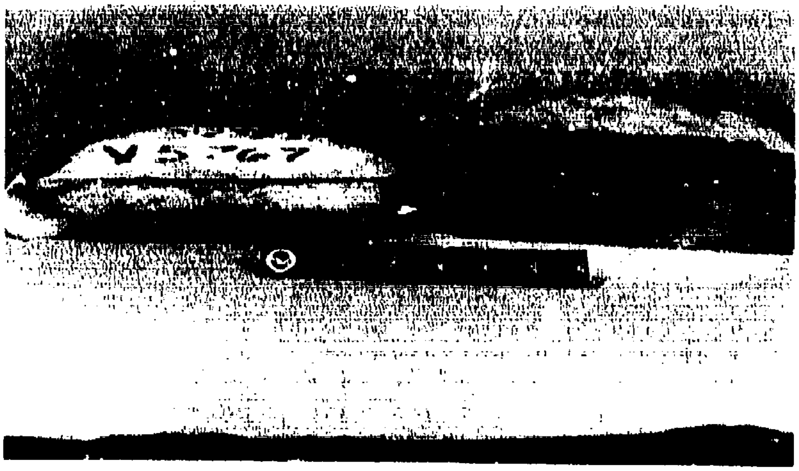


Figure 8

Isothermally forged TIMET gamma alloy ingot V-5767 (Ti-48Al-1V-.1[illegible]). Right, after first reduction; left, forging after second and final reduction. Note unhealed porosity on first upset.

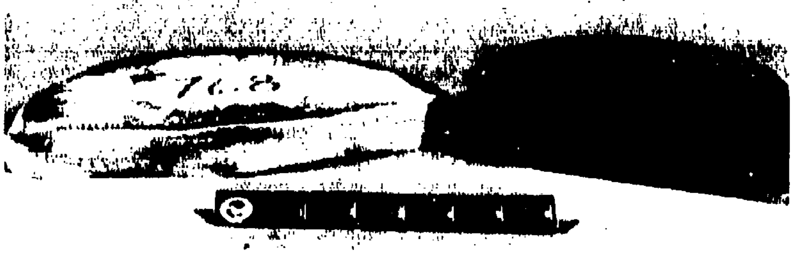


Figure 9

Isothermally forged TIMET gamma alloy ingot V-5768 (Ti-48Al-1V-.1[illegible]). Right, after first reduction; left, forging after second and final reduction.

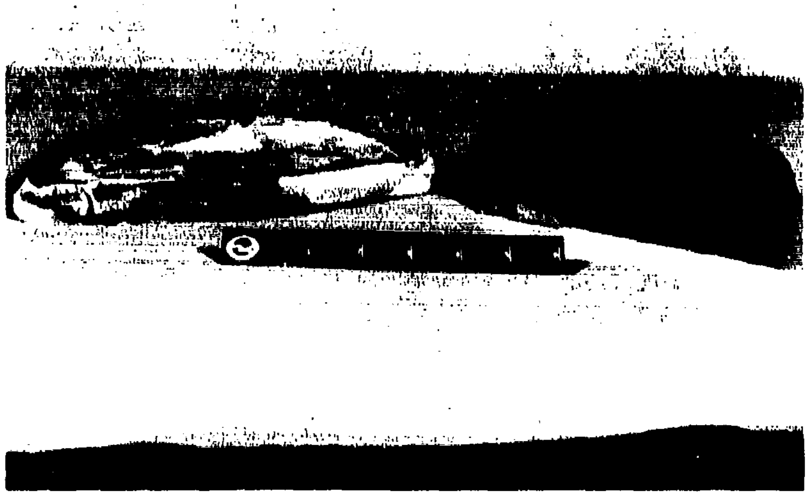


Figure 10

Isothermally forged TIMET gamma alloy ingot V-5769 (Ti-48Al-1V-.1C-.1O_2). Right, after first reduction; left forging after final reduction.

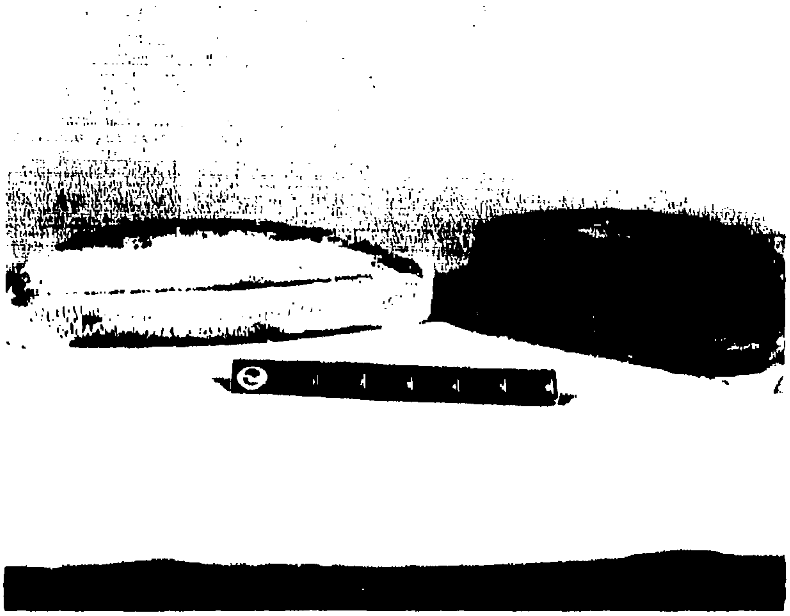


Figure 11

Isothermally forged TIMET gamma alloy ingot V-5770 (Ti-48Al-.5V-.1O_2). Right, after first reduction - note porosity; left forging after second and final reduction.

In the case of the gamma alloys, less was known about the heat treatment/microstructure effects since in the previous development program, the bulk of the data were for properties of as-forged material. Specimens of the three base alloys were heat treated at 55C (100F) intervals over the range of 982C (1800F) through 1425C (2600F). These were conducted using the platinum element tube furnace previously described. Heat treatment of specimen blanks was conducted in air except at temperatures over 1260C (2300F), where argon was used.

4. Task IV - Property Determination

All heat treated specimens of both types of alloys were examined metallographically. The Vickers hardness of each piece was also measured. Following this, three heat treatments were selected for three point bend testing at RT, 150C (300F) or 260C (500F). The bend specimens consisted of rectangular bars about 30 mm long x 5 mm wide x 2 mm thick (1.2 x .2 x .080 inches) cut and ground using 120 through 600 grit SiC papers. The supports and ram consisted of sapphire rods about 3.1 mm (.125 inch) in diameter. The loading rate was .025 cm/cm/sec (.01 in/in/min). If a specimen was deformed to full deflection without cracking, it was denoted as showing "plastic" behavior.

Based on the results of the above tests, a heat treatment was selected for each basic alloy composition for further mechanical property testing. At this time, the selected heat treatment for the Ti-25Al-14Nb baseline and Ti-48Al-1V were applied to forgings with varying interstitial contents. After the rectangular blanks were heat treated, tensile (notched and smooth), notched rupture and creep rupture specimens were machined using a combination of electrodischarge machining (EDM) and electrochemical grinding techniques. The specimen configurations selected are shown in Figure 12a-c. Except for the notched configuration, these specimens have been utilized in previous aluminide development programs.

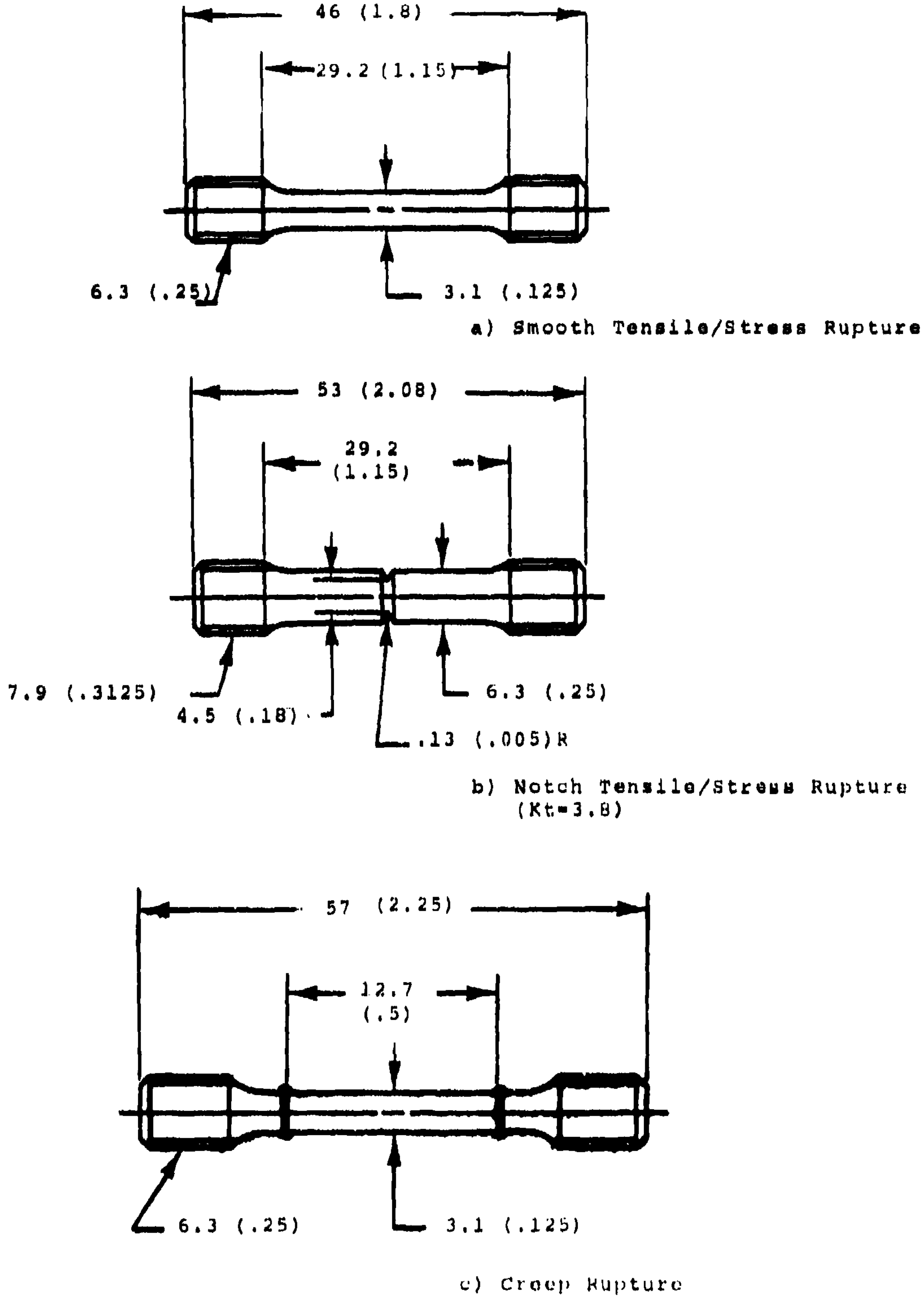


Figure 12

Mechanical property specimens used in this investigation. Dimensions are in mm (in.).

SECTION IV

RESULTS AND DISCUSSION

1. Alpha-Two Alloys

a. Heat Treatment Screening Trials

The primary objective of the alpha-two heat treatment study was to obtain a Widmanstätten type microstructure of very fine platelets within moderately large beta grains. All previous indications showed that this microstructure can provide the best balance of high temperature creep resistance and room temperature ductility. The average Vickers hardness aim was the 300-400 DPH range as previous studies have shown that the room temperature yield strength is about 1/3 of the Vickers number. Therefore, a range of 300-400 DPH would result in a room temperature yield strength around 695 MPa (100 ksi). In addition, it was decided to investigate an alpha two-beta quenched microstructure with higher hardness to see if useful ductility could be produced at higher strength levels. Aging treatments from 650C (1200F) to 815C (1500F) after solution treatment were selected for study.

The first step was to determine the beta transus temperatures of the various alloys using metallographic techniques. These are listed in Table 5. It was noted that the beta transus of the Ti-25Al-10Nb-4V composition was about 28C (50F) lower than the ternary Ti-25Al-14Nb composition, indicating that vanadium is a more potent beta stabilizer than niobium. The beta transus of the two Ti-25Al-14Nb MERL melted ingots (L and S) were both about 55C (100F) lower than the comparable TIMET composition. This was attributed to the lower oxygen content of the MERL ingots. It was at this point that it became evident that the attempt to produce a high oxygen content ingot had failed.

The entire series of heat treatments and resulting Vickers hardnesses are presented in Table 6. Metallographic examination was conducted on all specimens but, due to the large numbers, only those heat treatments/microstructures selected for further evaluation are illustrated (Figures 13-15). After reviewing the microstructure and Vickers hardness data, three heat treatments per alloy were selected for bend testing. These were beta annealing followed by salt quenching or air cooling plus aging, and alpha two-beta solution treating followed by oil quenching and aging, as listed in Table 7. Three point bend testing of the heat treated material was conducted at room temperature and 150C (300F), and the data are presented in Table 8. It can be seen that all specimens, including oil quenched, exhibited bend ductility in excess of 2.0% at room temperature. Many were in the 4-5% range. These values are

Table 5

Beta Transus Temperatures For
Alpha Two Alloy Compositions Selected
For Contract F33615-75-C-1167

Heat No.	Composition	Transus Temperature °C (°F)
V-5763	$Ti-25Al-14Nb-.1O_2$	Between 1135-1150 (2075-2100)
MERL "L"	$Ti-25Al-14Nb-0.025O_2$	Between 1080-1095 (1975-2000)
MERL "S"	$Ti-25Al-14Nb-.03O_2$	Between 1080-1095 (1975-2000)
V-5810	$Ti-25Al-10Nb-4V-.1O_2$	Between 1105-1120 (2025-2050)
V-5766	$Ti-24.5Al-13Nb-.1O_2$	Between 1120-1135 (2050-2075)

considerably higher than earlier results for the Ti-24Al-11Nb alloy(8). Material which had been quenched into 815C (1500F) salt had the highest ductilities for each composition. Rather than salt quench all alloys, however, it was decided to evaluate solution treating followed by air cooling and aging for the V-5766 composition since it was very similar to the baseline V-5763 material. While salt quenching appears to be the best procedure for avoiding the cooling rate/property sensitivity of the alpha two alloys, salt quenching may be impractical for very large structures or for applications where no subsequent surface machining is conducted, e.g., a blade or vane.

Material subjected to the selected treatments (one per alloy) was subsequently bend tested at 150C (300F). These data are also shown in Table 8 and indicate that little difference in ductility exists between the room temperature and 150C (300F) tests for the ternary Ti-Al-Nb (V-5763, V-5766) alloys. However, the V-5810 Ti-Al-Nb-V alloy exhibited a 30% increase in ductility over room temperature and could be plastically deformed to the limit of rig travel. This tends to verify previous data which indicated that vanadium additions seemed effective in improving intermediate temperature ductility.

The alpha-two-beta treatment, while yielding 2.3-2.8% room temperature bend ductility was dropped from further consideration because of the tendency to quench crack, and the ductilities measured were too low (we will see later that the tensile ductilities of beta processed material were not very high at ambient temperatures). The following three heat treatments were selected for tensile and creep-rupture evaluation:

Table 6
Average DPH Hardness Values For
Alpha-Two Alloys Heat Treated
For Contract F33615-75-C-1167

Alloy	Solution Temp. °C(°F)/Hr:	Cooling Method (1)	None	650(1200)/1/	705(1300)/1/	760(1400)/1/	760(1400)/8/	815(1500)/1/	815(1500)/8/
				Aging Temperature °C (°F)/Hrs (2)					
V-5763 Ti-25Al-14Nb-.1O$_2$	1120(2050)/1/	AC	275	269	270	272	-	-	-
		SAC	256	254	257	261	-	-	-
		OQ	365	-	451	368	-	358	-
	1160(2125)/1/	AC	317	-	-	302	302	298	319
		SAC	268	-	-	261	278	272	-
		FAC	432	-	-	455	-	392	356
		SQ	368	-	-	-	-	-	-
V-5766 Ti-24.5Al-13Nb-.1O$_2$	1107(2025)/1/	AC	248	254	258	250	-	-	-
		SAC	246	267	247	247	-	-	-
		OQ	363	-	489	416	-	-	412
	1150(2100)/1/	AC	302	-	-	308	283	298	315
		SAC	259	-	-	276	259	260	-
		FAC	416	-	-	430	-	388	-
		SQ	376	-	-	-	-	-	-
V-5810 Ti-25Al-10Nb-4V-.1O$_2$	1093(2000)/1/	AC	282	290	285	280	-	-	-
		SAC	276	273	275	276	-	-	-
		OQ	364	-	506	468	-	414	-
	1135(2075)/1/	AC	390	-	-	337	312	346	335
		SAC	294	-	-	282	274	291	-
		FAC	462	-	-	433	-	383	356
		SQ	363	-	-	-	-	-	-

(1) Air cool (AC) consisted of cooling in still air, open tray; slow air cooling (SAC) consisted of cooling the specimen inside a 2" titanium bar in still air; fan air cooling (FAC) consisted of cooling the specimen in an open tray in a stream of air. SQ indicates specimen was quenched into 815C (1500F) salt bath for 30 minutes.

(2) - means not evaluated.

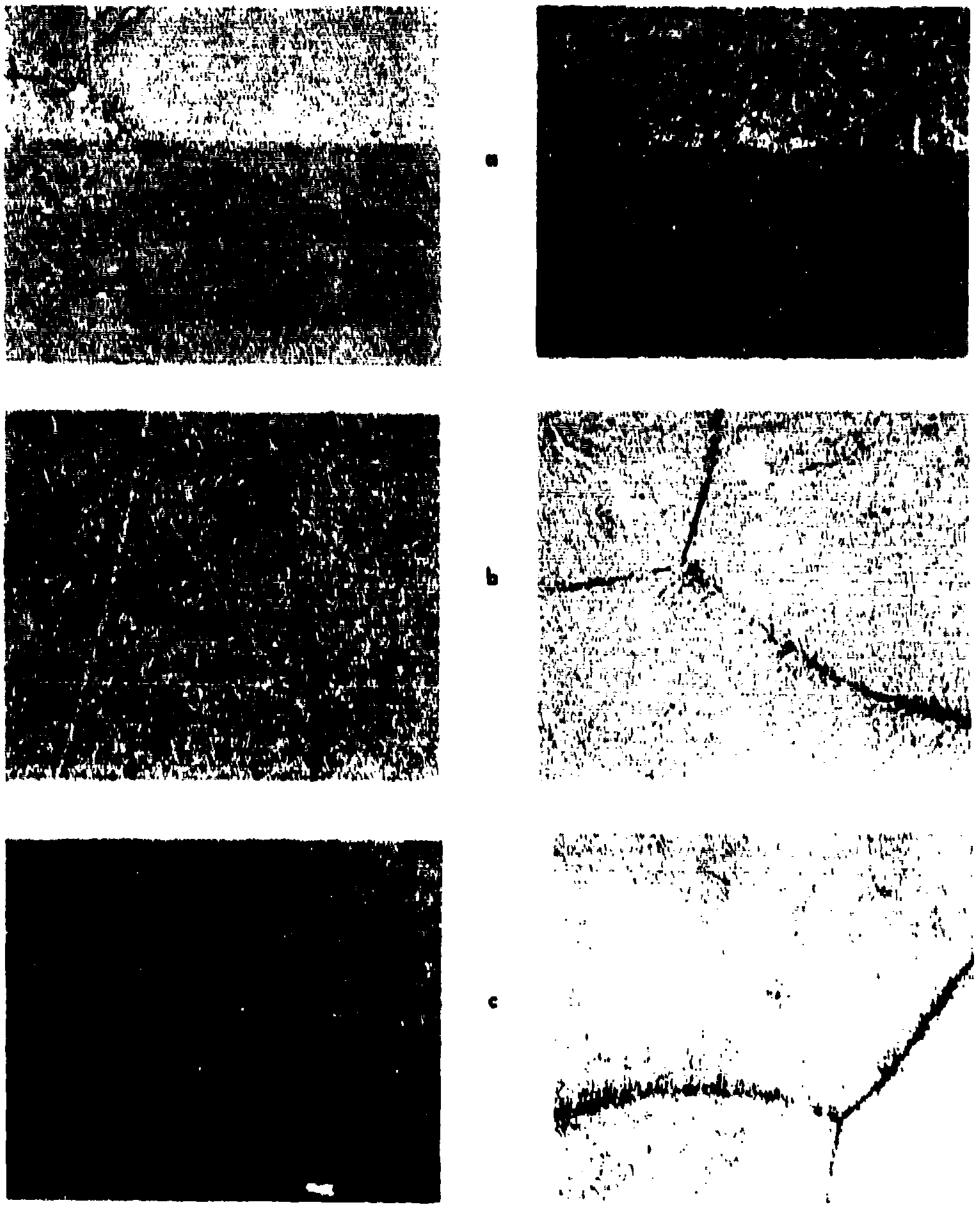


Figure 13 Mag: Left Column: 200X
Right Column: 500X

Microstructure of forged and heat treated alloy V-5763 (Ti-25Al-14Nb-.1O$_2$) selected for bend testing.

Row a) 1162 (2125)/1/SQ+815 (1500)/.5/AC
Row b) 1162 (2125)/1/AC + 815 (1500)/8/AC
Row c) 1162 (2125)/1/FAC + 815 (1500)/1/AC

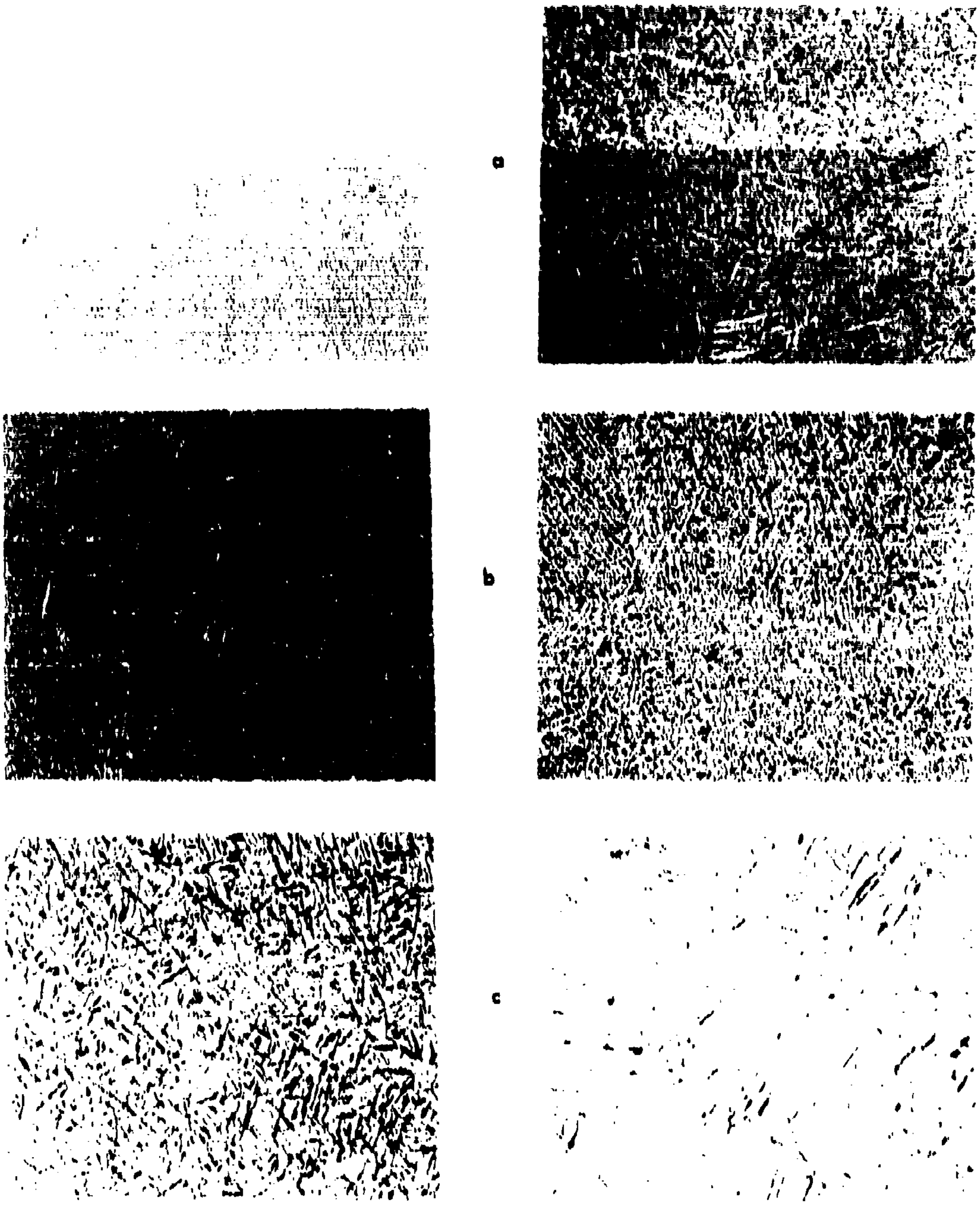


Figure 14 Mag: Left Column: 200X
Right Column: 500X

Microstructure of forged and heat treated alloy V-5766 (Ti-24.5Al-13Nb-.1O$_2$) selected for bend testing.

Row a) 1150 (2100)/1/SQ→815 (1500)/.5/AC
Row b) 1150 (2100)/1/AC + 815 (1500)/8/AC
Row c) 1095 (1975)/1/OQ + 760 (1400)/1/AC

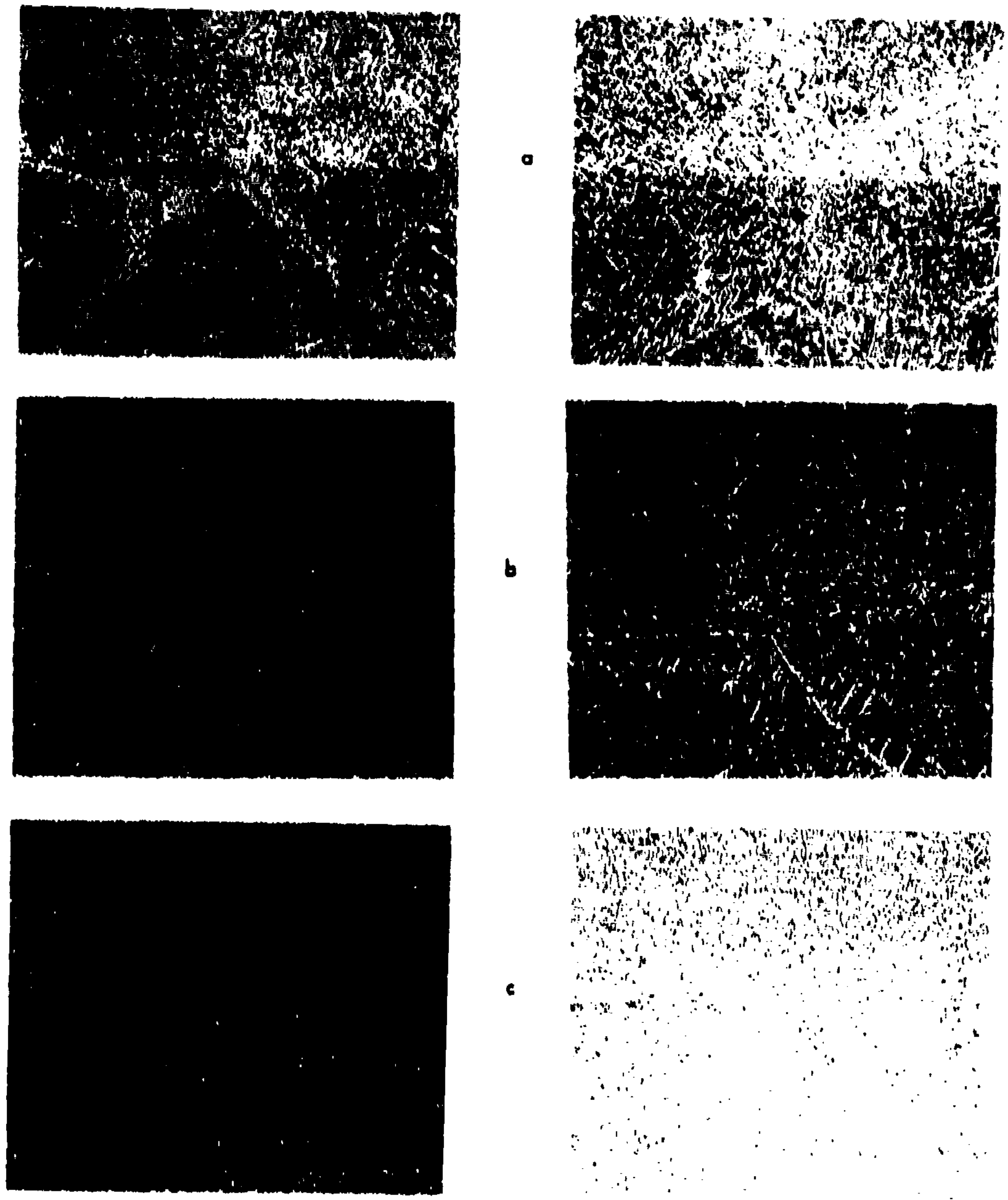


Figure 15

Mag: Left Column: 200X
Right Column: 500X

Microstructure of forged and heat treated alloy V-5810 (Ti-25Al-10Nb-4V-.1O_2) selected for bend testing.

Row a) 1135 (2075)/1/SQ→815 (1500)/.5/AC
Row b) 1135 (2075)/1/AC + 815 (1500)/8/AC
Row c) 1135 (2075)/1/FAC + 815 (1500)/1/AC

Table 7

Heat Treatments Selected
For Bend Test Evaluation
Of Alpha-Two Alloy Forgings

Alloy	Heat Treatments[1], °C(°F)	Avg. DPH Hardness
V-5763 ($Ti-25Al-14Nb-.1O_2$)	1162(2125)/1/SQ→815(1500)/.5/AC	368
	1162(2125)/1/AC + 815(1500)/8/AC	319
	1162(2125)/1/FAC + 815(1500)/1/AC	392
V-5766 ($Ti-24.5Al-13Nb-.1O_2$)	1150(2100)/1/SQ→815(1500)/.5/AC	376
	1150(2100)/1/AC + 815(1500)/8/AC	315
	1095[2](1975)/1/OQ + 760(1400)/1/AC	412
V-5810 ($Ti-25Al-10Nb--4V-.1O_2$)	1135(2075)/1/SQ→815(1500)/.5/AC	363
	1135(2075)/1/AC + 815(1500)/8/AC	335
	1135(2075)/1/FAC + 815(1500)/1/AC	356

(1) AC = Air cool, still air
FAC = Fan air cool
SQ = Salt quench
QO = Oil quench

(2) Adjusted to increase α_2 phase content.

Table 8

Bend Test Results For
Forged and Heat Treated Alpha-Two Alloys

Alloy	Heat Treatment, °C (°F)[1]	Test Temp. °C (°F)	0.2% Offset YS MPa,(Ksi)	Max Fiber Strength at Fracture, MPa (Ksi)	Max % Strain (Ductility)
V-5763 (Ti-25Al-14Nb-.1O_2)	1162(2125)/1/SQ+815(1500)/.5/AC	RT	1140 (165.4)	1570 (227.7)	4.06
		RT	1148 (166.5)	1680 (243.5)	5.16
		150 (300)	1171 (169.9)	1581 (229.3)	4.21
	1162(2125)/1/AC + 815(1500)/8/AC	RT	1270 (184.0)	1664 (241.4)	3.84
		RT	1217 (176.5)	1666 (241.7)	3.94
	1162(2125)/1/FAC + 815(1500)/1/AC	RT	1444 (209.4)	1767 (256.3)	2.89
		RT	1522 (220.8)	1898 (275.4)	3.28
V-5766 (Ti-24.5Al-13Nb-.1O_2)	1150(2100)/1/SQ+815(1500)/.5/AC	RT	1350 (195.8)	1844 (267.5)	4.89
		RT	1191 (172.8)	1675 (243.0)	4.22
	1150(2100)/1/AC + 815(1500)/8/AC	RT	1036(150.3)	1511 (219.3)	3.74
		RT	1137 (164.9)	1552 (225.1)	3.44
		150 (300)	1028 (149.1)	1456 (211.2)	4.21
	1095(1975)/1/OQ + 760(1400)/1/AC	RT	1597 (231.6)	1860 (270.4)	2.31
		RT	1557 (225.8)	1975 (286.5)	2.82
V-5810 (Ti-25Al-10Nb-4V-0.1O_2)	1135(2075)/1/SQ+815(1500)/.5/AC	RT	1213 (175.6)	1638 (237.6)	4.02
		RT	1311 (190.1)	1580 (229.0)	2.83
		150 (300)	1116 (161.8)	1628 (236.1)	4.93 (Plastic)[2]
	1135(2075)/1/AC + 815(1500)/8/AC	RT	916 (132.9)	1190 (172.5)	2.64
		RT	838 (121.5)	1242 (180.2)	3.21
	1135(2075)/1/FAC + 815(1500)/1/AC	RT	1338 (194.0)	1684 (244.2)	3.42
		RT	1300 (188.4)	1561 (226.4)	2.71

(1) AC = Air cool, still air
FAC = Fan air cool
SQ = Salt quench
OQ = Oil quench

(2) "Plastic" behavior indicates no cracking at the limit of ram travel.

Heat No.	Composition	Heat Treatment °C (°F)
V-5763	Ti-25Al-14Nb-.1O$_2$	1162(2125)/1/SQ→815(1500)/.5/AC
V-5766	Ti-24.5Al-13Nb-.1O$_2$	1150(2100)/1/AC + 815(1500)/8/AC
V-5810	Ti-25Al-10Nb-4V-.1O$_2$	1135(2075)/1/SQ→815(1500)/.5/AC

Rectangular blanks, about 75 mm (3 inches) long x 40 mm (1.5 inches) square, were cut from the pancake forgings for heat treatment to provide the necessary specimens. The smaller MERL ingots were blanked and heat treated in a similar manner as V-5763.

b. Tensile Testing

Tensile data for the alpha-two alloys are presented in Table 9 for specimens fabricated by Pratt & Whitney Aircraft and tested by the Air Force Materials Laboratory. The results may be summarized as follows:

- Room temperature yield and ultimate tensile strengths of the three major compositions (TIMET material) were not significantly different, although the vanadium containing alloy was about 70 MPa (10 ksi) lower in strength than the ternary alloys. All alloys were stronger than the 700 MPa (100 ksi) yield strength goal.

- Room temperature percent elongation of the ternary alloys was about one percent higher (2.1-2.8%) than the 1-8-2.0% values of the vanadium containing alloy, but reduction in area was about a percent higher in the case of the latter alloy.

- Both strength and ductility tended to peak at 427C/800F. This has been observed in previous studies[9].

- The lower oxygen content of the MERL ingots "L" and "S" resulted in 140-210 MPa (20-30 ksi) lower 0.2% yield strengths than V-5763 at room temperature. The similarity of the ultimate strengths are due to the ductility differences. At higher temperatures, the yield strength differences are still evident.

- The lower oxygen content MERL ingots "L" and "S" showed a significant 2-3:1 increase in room temperature and 260C (500F) tensile ductility compared to the (V-5763) baseline material; at higher temperatures, there was no discernible effect. These data indicate that the selection of high purity titanium sponge can play a large role in improving ambient and intermediate temperature ductility in aluminide alloys.

Table 9

Smooth Tensile Properties of
Forged and Heat Treated Alpha-
Two Titanium Aluminide Alloys

Heat/Alloy	Heat Treatment °C (°F)/Hrs/	Test Temp. °C (°F)	0.2% Yield Strength MPa	(Ksi)	Fracture or Ultimate Strength MPa	(Ksi)	% Elong.	% Reduction in Area
V-5763	1162(2125)/1/SQ 815(1500)/.5/AC	RT	830	(120.4)	976	(141.6)	2.1	2.3
(Ti-25Al-14Nb-.1O_2)		RT	796	(115.5)	922	(133.7)	2.6	2.4
		260 (500)	620	(89.9)	969	(140.5)	5.8	13.5
		427 (800)	583	(84.5)	1013	(146.9)	14.1	15.3
		538 (1000)	635	(92.1)	831	(120.5)	7.9	27.9
MERL Ingot "S"	1107(2025)/1/SQ 815(1500)/.5/AC	RT	660	(95.7)	932	(133.7)	5.7	6.3
(Ti-25Al-14Nb-.03O_2)		RT	629	(91.3)	936	(135.7)	5.8	7.8
		260 (500)	522	(75.7)	1131	(164.1)	20.4	22.1
		538 (1000)	481	(69.8)	895	(129.8)	9.7	20.9
MERL Ingot "L"	1107(2025)/1/SQ 815(1500)/.5/AC	RT	613	(88.9)	908	(131.7)	6.9	7.2
(Ti-25Al-14Nb-.025O_2)		RT	686	(99.5)	1007	(146.1)	7.0	7.1
		260 (500)	525	(76.1)	1163	(168.7)	13.8	12.9
		538 (1000)	523	(75.8)	916	(132.9)	6.5	21.6
		650 (1200)	507	(73.6)	774	(112.3)	6.8	17.3
V-5810	1135(2075)/1/SQ 815(1500)/.5/AC	RT	721	(104.6)	873	(126.6)	1.7	3.2
(Ti-25Al-10Nb-4V-0.1O_2)		RT	725	(105.2)	880	(127.7)	2.0	3.7
		260 (500)	558	(81.0)	930	(134.9)	8.1	10.9
		427 (800)	558	(80.9)	1068	(154.9)	14.9	16.4
		538 (1000)	525	(76.1)	985	(142.8)	9.2	22.3
V-5766	1150(2100)/1/AC + 815(1500)/8/AC	RT	785	(113.8)	948	(137.5)	2.8	2.2
(Ti-24.5Al-13Nb-.01O_2)		260 (500)	606	(87.9)	982	(142.2)	7.4	13.9
		427 (800)	556	(80.7)	1016	(147.4)	12.1	19.1
		538 (1000)	557	(80.8)	868	(125.9)	7.4	14.8

Notched tensile test data are presented in Table 10. The baseline (V-5763) specimens had an average notched tensile strength of 648 MPa (94 ksi) which gave a notched/smooth strength ratio of 0.7. Specimens from the low oxygen content material (MERL ingot "L") had an average notched strength of 766 MPa (111.6 ksi). The notch/smooth ratio in low oxygen material was 0.8, reflecting the somewhat higher ductility. Notched tensile strength of V-5763 material which had been thermally charged with 200-235 ppm of hydrogen prior to heat treatment had an average notch strength of 770 MPa (111.7 ksi).

c. Room Temperature Notched Stress Rupture

These tests were conducted by step loading notched tensile specimens which is a standard method of evaluating hydrogen effects. The initial stress was 414 MPa (60 ksi) and this was increased 70 MPa (10 ksi) every five hours. Testing was limited to baseline (V-5763) material in the as-received (70-90 ppm H_2) and thermally hydrogenated (200-235 ppm H_2) conditions. Test results, shown in Table 11, revealed that increased hydrogen content reduced rupture life. Baseline tests ran 30-40 hours with a final rupture stress of 828-965 MPa (120-160 ksi), while the hydrogen charged specimens ruptured in 25 hours at a stress of 690 MPa (100 ksi).

d. Creep Rupture

Creep rupture tests were conducted using specimens from the three major TIMET compositions. Interstitial variable material was not tested since no previous indications of strong oxygen effects at elevated temperature had been observed. Tests were conducted over the potential application range of the alpha two alloys, i.e., from 593C (1100F) to 815C (1500F). The data showing time to 1% creep extension and rupture are given in Table 12 and presented on a Larson-Miller format in Figure 16. There is no major difference in time to rupture lives for the three alloys although the vanadium containing (V-5810) material appears to be about 14C (28F) lower in rupture capability. The same margin also existed in time to one percent creep extension. For comparison purposes, various other alloys were plotted on the Larson-Miller graph. These included beta annealed Ti-6Al-2Sn-4Zr-2Mo (Ti-6242) which has the highest creep strenth of the commercially used conventional titanium alloys, and two cast nickel base alloys, Waspaloy and Inconel 713C. It can be seen that the alpha two compositions exhibited nearly a 110C (200F) advantage in life to 1% creep and to rupture over the Ti-6242. In comparison to the nickel alloys, the alpha two compositions were nearly equivalent to Waspaloy but about 55C (100F) lower in capability than the Inconel 713C plotted on a density corrected basis.

Table 10

Effect of Interstitial Elements on
Room Temperature Notched (Kt=3.9)
Tensile Strength of Ti-25Al-14Nb Alloy

Heat/ Identity	Condition	Interstitial Content, ppm H_2	O_2	Tensile Str. MPa (Ksi)	Average Tensile Str. MPa (Ksi)
V-5763	Baseline[1]	70-90	760	583 (84.6)	648 (94.0)
V-5763	Baseline	70-90	760	712 (103.3)	
MERL "L"	High O_2[2]	70-90	300	762 (110.5)	766 (111.6)
MERL "L"	High O_2	70-90	300	771 (111.8)	
V-5763	Baseline + H_2[3]	200-235	760	811 (117.6)	770 (111.7)
V-5763	Baseline + H_2	200-235	760	729 (105.7)	

(1) Specimens machined from conventionally forged pancakes and heat treated at 1160C (2125F)/1/salt quench to 815C (1500F)/.5/AC.

(2) Isothermally forged laboratory scale pancake given baseline heat treatment. An unsuccessful attempt was made to get 1500 ppm oxygen.

(3) Specimens machined as in (1), charged in Sievert's apparatus, then given baseline heat treatment.

Table 11

Room Temperature Notched (Kt=3.9)
Stress-Rupture Strength of
Alloy V-5763 (Ti-25Al-14Nb)

Condition	Hours To Rupture	Final Stress[3] MPa (Ksi)
Baseline[1]	30.4	828 (120)
Baseline	40.1	965 (140)
H_2 Charged[2]	25.0	690 (100)
H_2 Charged	25.0	690 (100)

(1) Baseline Heat Treatment - 1160C (2125F/1/salt quench
815C (1500F)/.5/AC

(2) Charged to 200-235 ppm H_2 in Sievert's apparatus, then given baseline heat treatment.

(3) Initially loaded at 414 MPa (60 ksi), load increased 70 MPa (10 ksi) every five hours.

Table 12

Creep-Rupture Test Results
For Heat Treated Alpha-Two Alloys

Alloy	Heat Treatment °C (°F)	Test Condition °C/MPa (°F/Ksi)	Time, Hours 1%	Rupture	Post Test %El	%RA
V-5763 (Ti-25Al-14Nb-.1O_2)	1162(2125)/1/SQ 815(1500)/.5/AC	593/414 (1100/60)	48.6	199.9	15.1	3.7
		650/380 (1200/55)	10.7	77.5	8.5	6.7
		650/380 (1200/55)	26.5	41.4	7.3	6.8
		704/275 (1300/40)	3.5	26.6	(1)	22.7
		760/207 (1400/30)	2.0	19.7	(1)	33.1
		815/103 (1500/15)	6.0	95.7	(1)	83.6
V-5810 (Ti-25Al-10Nb-4V-.1O_2)	1135(2075)/1/SQ 815(1500)/.5/AC	593/414 (1100/60)	13.1	204.1	(1)	6.3
		650/380 (1200/55)	4.3	50.4	(1)	11.3
		650/380 (1200/55)	5.1	68.4	(1)	9.3
		704/275 (1300/40)	3.0	32.6	(1)	2.1
		760/207 (1400/30)	1.6	20.3	(1)	18.7
		815/103 (1500/15)	3.8	65.0	(1)	20.7
V-5766 (Ti-24.5Al-13Nb-.1O_2)	1150(2100)/1/AC + 815(1500)/1/AC	593/414 (1100/60)	30.2	73.7	5.9	1.9
		650/380 (1200/55)	5.5	60.6	10.6	9.8
		650/380 (1200/55)	7.0	78.7	(1)	6.3
		704/275 (1300/40)	4.0	40.4	15.7	14.4
		760/207 (1400/30)	2.5	31.2	20.6	22.8
		815/103 (1500/15)	7.2	156.0	(1)	67.0

(1) Cannot measure elongation due to nature of fracture surfaces.

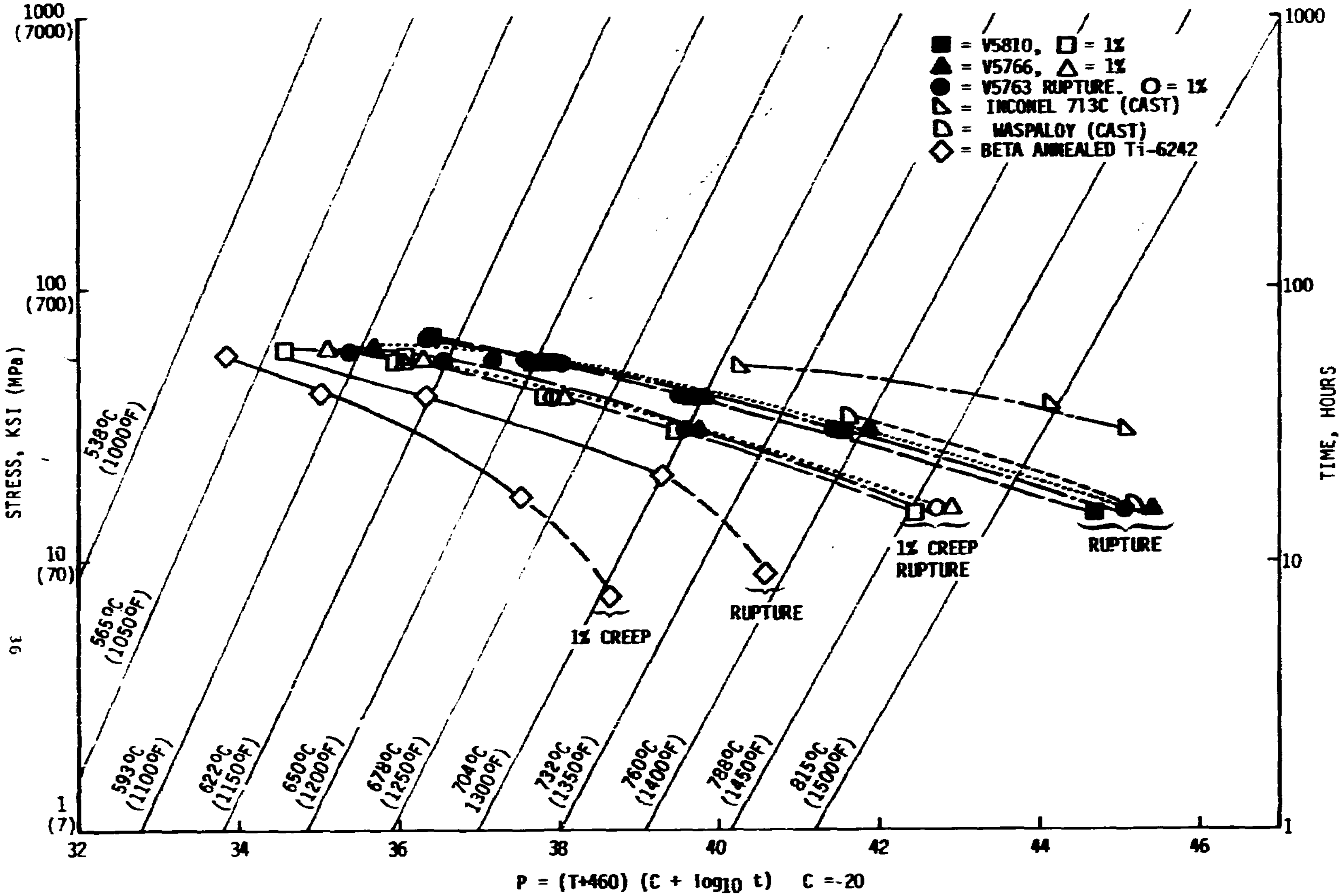


FIGURE 16

Time to 1% creep elongation and rupture for forged and heat treated alpha-two alloys compared to various titanium and nickel base alloys.

This report summarizes the final year of the alloy development effort initiated in 1975 at which time certain property goals were set. Simply stated, these were tensile properties equivalent to beta processed Ti-6242 and density corrected creep rupture properties equal to IN713C. Low oxygen alloys of the type studied during the current year meet the first goal, but the creep properties fall short. However, especially at higher temperatures, the property levels measured represent a significant improvement over the Ti-24Al-11Nb alloys studied in earlier years and are adequate for many of the potential applications of this alloy type. Interstitial element effects parallel quite closely the behavior pattern of conventional alpha beta titanium alloys. High hydrogen levels cause premature crack nucleation and growth in severely notched locations; experience has shown that such behavior is usually accompanied by low lives under cyclic stresses. Such undesirable effects are typically controlled by the enforcement of tight specification limits. Oxygen additions also have similar effects to that observed in alpha beta alloys. High levels increase strength at the expense of low temperature ductility, again, a tendency paralleled by conventional alloys. Although the strength differential decreases at intermediate temperatures, it does persist over the range of test temperatures studied.

In summary, the alloys of the type Ti-25Al-13.5Nb exhibit mechanical properties which indicate that a useful engineering material has been identified. Oxygen contents of $\sim$0.05 could improve ambient temperature ductility properties and thus aid general handleability. It is clear that niobium may be replaced by vanadium, at least to the 4% level, with no sacrifice in mechanical properties. The limits of such replacement are not known and further, the effect of quinary additions such as molybdenum, a yet more potent beta stabilizer, would also be of interest but were beyond the scope of the present program. Considerable additional information is needed to build on the present base before components can be designed and engine tested. Present contractual efforts will go a long way toward fulfilling these requirements by generating the property information needed for design purposes.

2. Gamma Alloys

a. Heat Treatment Screening Trials

Previous alloy development studies involving gamma titanium aluminide alloys have concentrated on material in the as-forged condition due to the limited availability of material and large numbers of compositions evaluated. Thus, less was known about the heat treatment/microstructure/property response of these alloys at the outset of the present work.

A brief study conducted on a forged and heat treated Ti-50Al-1W composition(6) showed that forging temperature had a strong influence on tensile and yield strength but little effect on ductility. Forging at 1010C (1850F) resulted in a 20-30% increase in strength compared to forging at 1093C (2000F). (The higher temperature had been used on all previous work.) Solution treating at 1204C (2200F) and aging at 815C (1500F) lowered strength, increased ductility and eliminated most of the strength variations resulting from the different forging temperatures. Creep rupture life of heat treated material increased 3-4 fold over that of forged material and the largest increase was measured for material forged at the lower temperature of 1010C (1850F). Direct aging at 815C (1500F) lowered strength slightly from the forged condition and preserved the differences between the forging temperatures. Creep rupture lives were poor, however, compared to those on the forged or solution treated and aged condition.

Other studies conducted at Pratt & Whitney and the United States Air Force indicated that solution treating at very high temperatures approaching 1370C (2500F) resulted in an acicular microstructure in most alloys. A significant increase in creep resistance for such structures was measured, but there was a loss in tensile ductility at temperatures up to about 538C (1000F).

Using these available data as a starting point, test coupons from the three major composition variations (V-5767, 5769, 5770) were solution treated at 55C (100F) intervals over the range 982-1425C (1800-2600F). While not all of the microstructures can be shown due to space limitations, the behavior of the three compositions was similar. Photomicrographs depicting typical microstructures are shown in Figures 17(a-g). It can be seen that, in the range 982C (1800F) to 1260C (2300F), the microstructure consisted predominantly of an equiaxed gamma phase, but the grain size increased from an ASTM value of 7-8 to

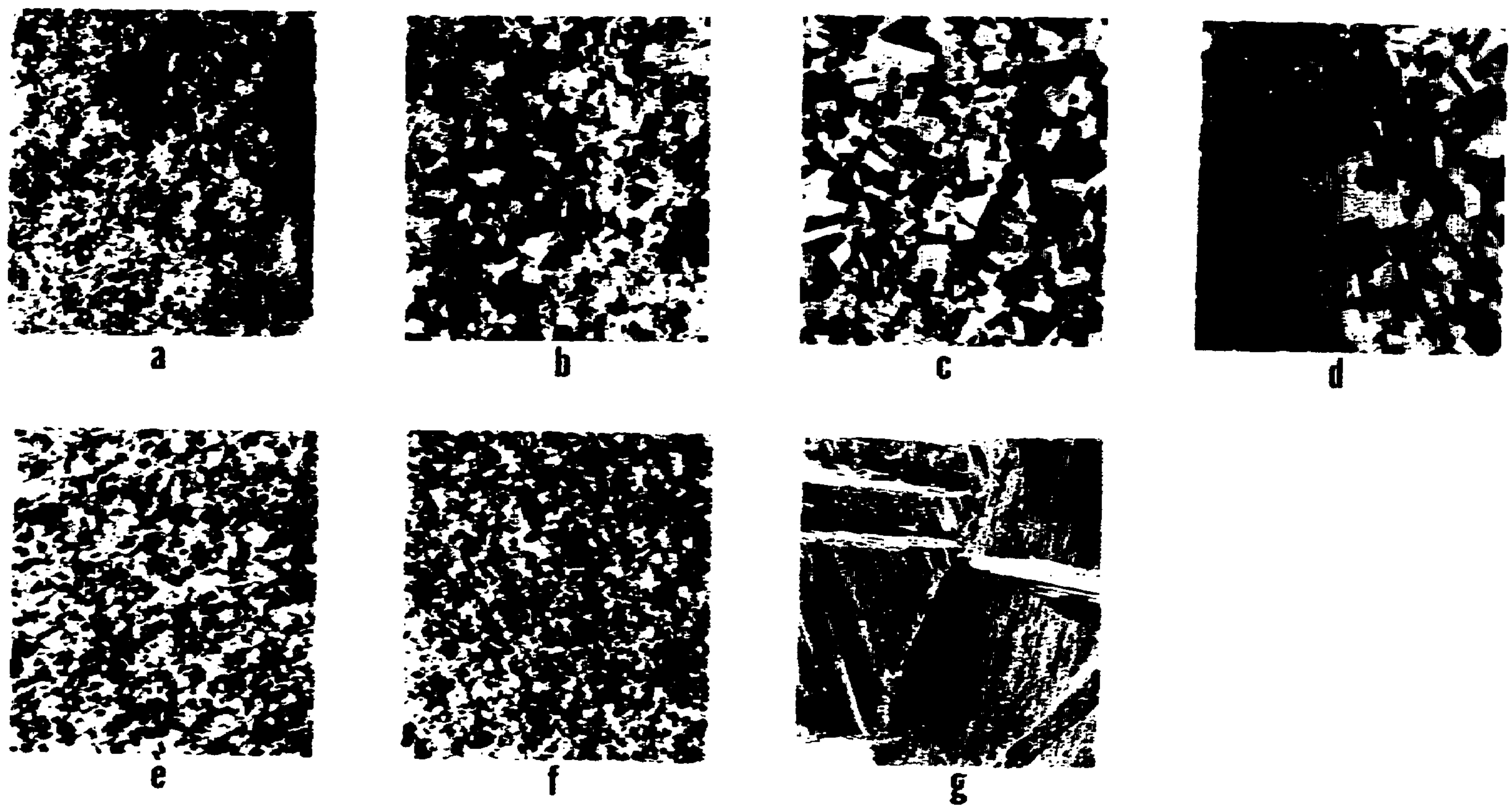


Figure 17

Effect of solution temperature on the microstructure of Ti-48Al-1V-0.1C, (V-5767). a) 982C (1800F); b) 1150C (2100F); c) 1204C (2200F); d) 1260C (2300F); e) 1315C (2400F); f) 1370C (2500F); g) 1425C (2600F). The samples were heat treated in argon for one hour and cooled at a rate equivalent to air cooling.

ASTM 3-4. A small amount of alpha-two phase was also present, becoming more readily detectable as the grain size coarsened. At 1315C (2400F). all alloys showed a surprising reversion to a fine grain size in the ASTM 7-8 range, although with occasional areas of coarser 5-6. At 1370C (2500F), this phenomena was still apparent in V-5767 (Ti-48Al-1V-.1$0_2$) and V-5770 (Ti-48Al-0.5V-.1$0_2$) while alloys V-5768 (Ti-48Al-1V-.13$0_2$) and V-5769 (Ti-48Al-1V-.1C-.1$0_2$) exhibited full acicular microstructures. At 1425C (2600F), all alloys were fully acicular. At first, it was difficult to understand how a grain size could become smaller as solution temperature increased instead of a gradual increase until a full acicularity occurred. A transmission electron microscopy (TEM) study was conducted, and it is now considered that a kinetic effect may have been involved. Figure 18 shows the major changes found in the thin foil study which may be summarized as follows. At temperatures up to 2300F, a two phase structure is evident with little or no substructure in the alpha two phase (Figure 18a). When the solution treatment temperature reaches 1315C (2400F), the character of second phase regions become two phase with a lamellar structure. It is probable that this results from transformation of the beta phase during cooling from the solution treatment temperature. Thus, we may conclude that the alpha-two phase is only a minor impediment to grain growth but that the beta phase is more effective in pinning grain boundaries. As the solution treatment temperature is increased above 1315C (2400F), the amount of beta phase increases until, at 1425C (2600F), this is the only constituent present. On cooling, the phase transforms to the lamellar alpha two/gamma structure (Figure18b) although the precise features of this transformation are unclear.

The apparent grain size anomaly, noted above, may now be explained as follows. If a specimen is inserted into a preheated furnace, formation of the beta phase may be sufficiently rapid to prevent substantial grain growth. This has been proved by a subsequent experiment performed as follows. Samples of V-5767 (Ti-48Al-1V-.1$0_2$) were heated to 1150C (2100F), and equilibrated, and heated slowly at 110C (200F) hour to 2500F. The resulting microstructure (Figure 19) was equiaxed and coarse grained with an ASTM grain size of 4. By heating slowly, it was apparent that the gamma phase started coarsening and the subsequent formation of the beta phase was too late to prevent grain growth.

Based on these studies, specific heat treatments were selected, applied to specimens and evaluated. Vickers hardness data measured after the treatments are shown in Table 13. Hardness did not vary much with heat treatment values being virtually independent of solution and/or aging temperature, at least until the beta transus was exceeded. The specimens with an acicular microstructure were about 20-30 points harder than

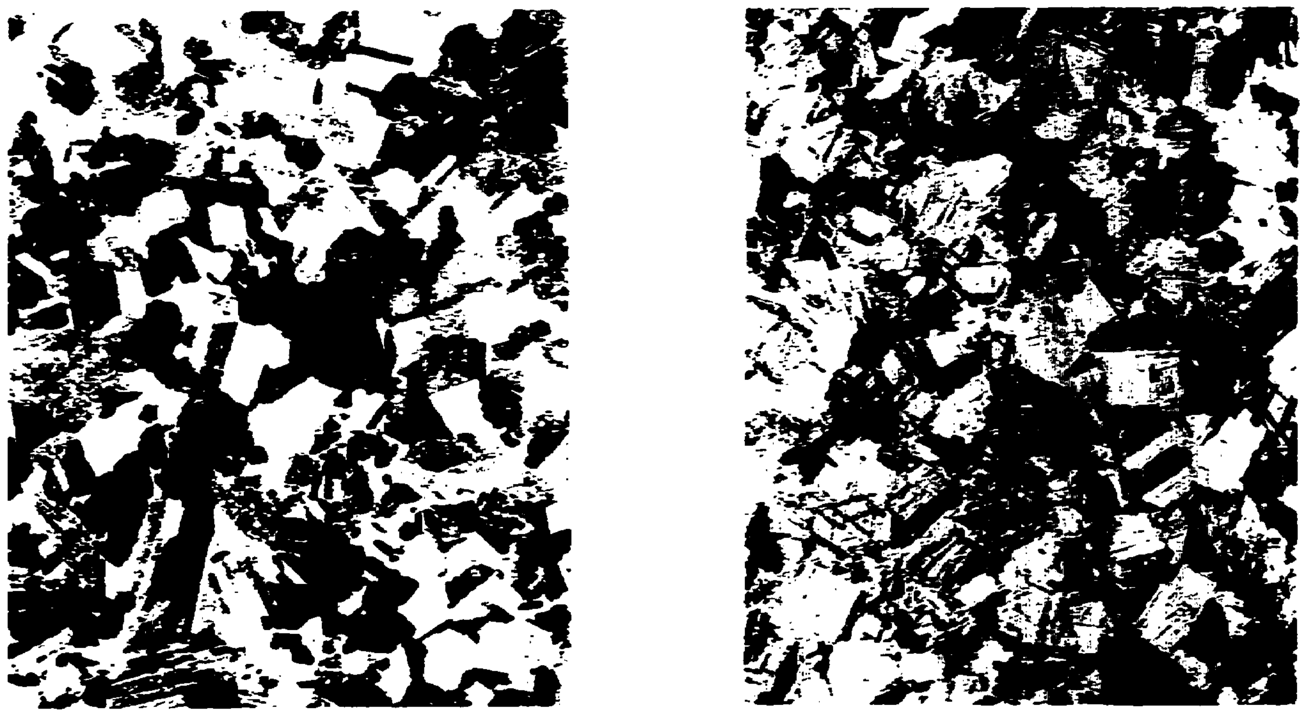
Figure 16

Transmission electron micrographs showing the variation in structure with solution temperature of Ti-48Al-1V (V-5767).

a) 1150C (2100F)/1/water quench; alpha-two (arrow) in gamma

b) 1315C (2400F)/1/water quench; gamma plus transformed beta (bottom)

c) 1425C (2600F)/1/water quench; fully transformed beta



(a) (b)

Figure 19 Mag: 100X

Microstructure of Ti-48Al-1V (V-5767) alloy heated to 1150C (2500F) at 110C (200F)/hour and furnace cooled (left) or cooled in flowing argon (right). Note coarser grain size compared to V-5767 heated rapidly to 1370C (2500F) shown in Figure 17(f).

Table 13

Average DPH Hardness Values for Baseline Gamma Alloys Heat Treated for Contract F33615-75-C-1167

Solution Temp.* °C(°F)/Hrs.	Aging Temp. °C(°F)/Hrs.	V-5767 (Ti-48Al-1V-.1O$_2$)	V-5769 (Ti-48Al-1V-.1C-.1O$_2$)	V-5770 (Ti-48Al-.5V-.1O$_2$)
-	760(1400)/8/AC	210	289	220
-	815(1500)/8/AC	211	285	222
-	871(1600)/8/AC	213	283	220
982(1800)/1/	-	212	279	215
1093(2000)/1/	-	203	277	190
1150(2100)/1/	-	205	266	193
1150(2100)/1/	815(1500)/8/AC	183	260	-
1204(2200)/1/	-	208	256	197
1204(2200)/1/	815(1500)/8/AC	195	260	-
1260(2300)/1/	-	208	247	199
1260(2300)/1/	815(1500)/8/AC	193	253	-
1315(2400)/1/	-	206	268	203
1315(2400)/1/	815(1500)/8/AC	200	270	-
1370(2500)/1/	-	222	293	209
1370(2500)/1/	815(1500)/8/AC	219	313	-
1425(2600)/1/	-	228	318	241
1425(2600)/1/	815(1500)/8/AC	231	325	-

*Solution treatments conducted in argon atmosphere, cooled in flowing argon at a rate equivalent to air cooling.

equiaxed structures. In most cases, aging at 815C (1500F) caused hardness to increase slightly. After reviewing the metallography and hardness results, which are listed in Table 14, three heat treatments were selected for bend testing. The objectives were to test a fine grain equiaxed structure (ASTM 9), a coarse grain (ASTM 3-5) equiaxed structure and a structure with transformed beta phase present, although it was felt that a fully acicular, very large grain size microstructure would probably be too brittle for practical consideration. To achieve a partially acicular microstructure, the bend specimens were heat treated 28C (50F) below the limit of the single phase beta field. The resulting microstructures of the heat treated bend specimens are shown in Figures 20-22 and room temperature and 260C (500F) bend test data are presented in Table 15.

The range of bend ductility was from 1-2% and showed only minor variations with heat treatment for all three alloys. Some observations can be made, however. First, the ductility of solution heat treated specimens was slightly higher than as-forged specimens in all cases. Second, the ductility of the as-forged low oxygen content Ti-48Al-1V ingot was nearly 50% higher than the baseline V-5767 composition. Third, the V-5769 carbon containing Ti-48Al-1V composition ahd the best ductility of all compositions at nearly twice the strength level.

Selection of heat treatments for tensile and creep rupture testing covered a slightly different range of heat treatments for the various compositions. This was done partly to expand our knowledge of heat treatment/microstructure/property response in gamma systems. The base composition, Ti-48Al-1V-.1O_2 (V-5767), the alloy Ti-48Al-1V-.13O_2 (V-5768) and the alloy Ti-48Al-1V-.03O_2 (MERL melt) were solution treated at 1150C (2100F) followed by aging at 815C (1500F). The Ti-48Al-0.5V-0.1O_2 (V-5770) composition was directly aged at 815C (1500F), while the V-5769 carbon containing composition was given high temperature solution treatment at 1315C (2400F) followed by an 815C (1500F) age to achieve a microstructure with some acicular phase.

b. Tensile Testing

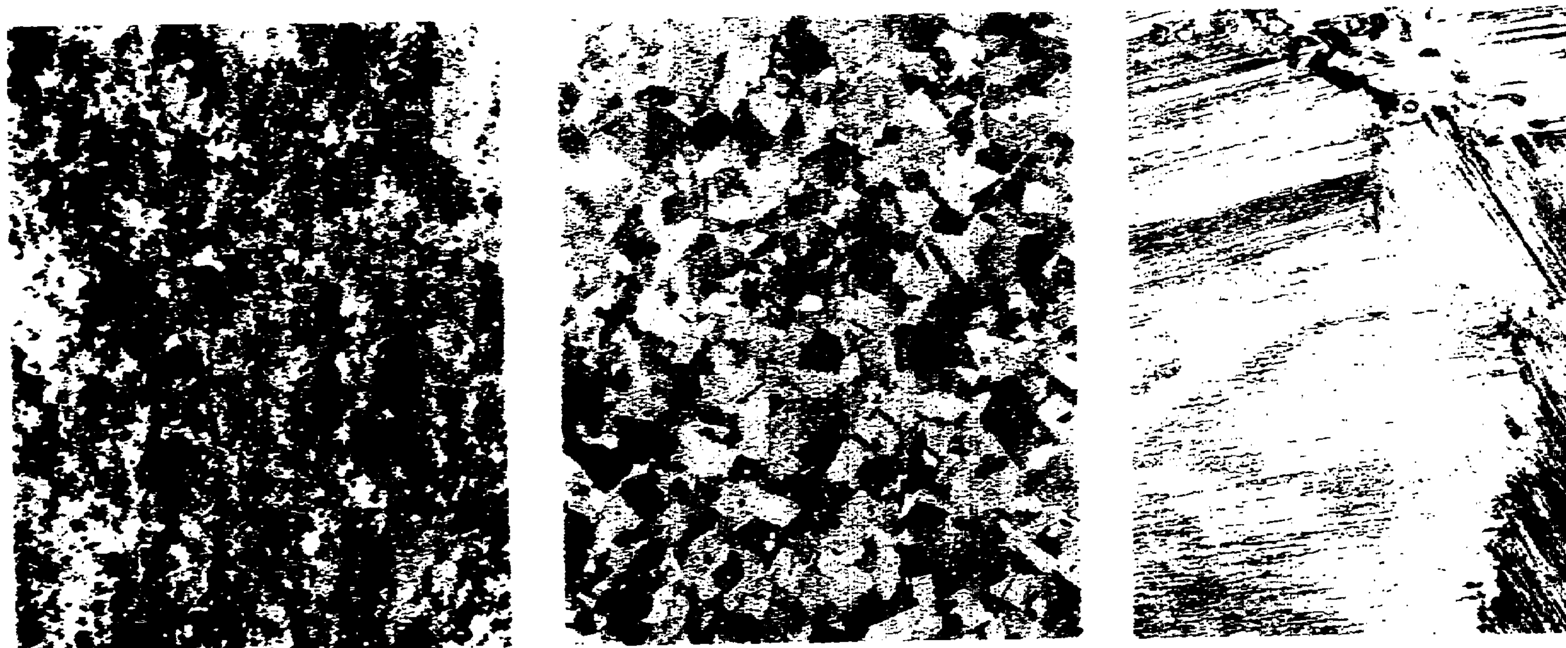
Smooth tensile data are given in Table 16. As before, the specimens were supplied by Pratt & Whitney Aircraft and tested at the Air Force Materials Laboratory. The following conclusions were reached after studying the test data:

- The Ti-48Al-1V-.1O_2 (V-5767) composition was low in strength and had only about half of the ductility previously measured on laboratory scale as-forged pancakes.

Table 14

Heat Treatments Selected For
Bend Test Evaluation of
F33615-75-C-1167 Gamma Alloys

Alloy	Heat Treatment °C (°F)/Hrs.	DPH Hardness	Microstructure
V-5767 (Ti-48Al-1V-.1O$_2$)	815(1500)/8/AC	220	Equiaxed grains ASTM 8-9
	1150(2100)/1/Argon Cool + 815(1500)/8/AC	189	Equiaxed grains ASTM 5-6
	1398(2550)/1/Argon Cool + 815(1500)/8/AC	228	Duplex grains; ASTM 2-3 containing an acicular phase.
V-5769 (Ti-48Al-1V-.1C-.1O$_2$)	815(1500)/8/AC	295	Equiaxed grains ASTM 9-10
	1150(2100)/1/Argon Cool + 815(1500)/8/AC	268	Duplex structure equiaxed
	1315(2400)/1/Argon Cool + 815(1500)/8/AC	245	Mixture of equiaxed and acicular grains ASTM 7-9
V-5770 (Ti-48.5Al-.5V-.1O$_2$)	815(1500)/8/AC	219	Equiaxed grains ASTM 9 (occ. 7)
	1150(2100)/1/Argon Cool + 815(1500)/8/AC	197	Equiaxed grains ASTM 6
	1370(2500)/1/Argon Cool + 815(1500)/8/AC	224	Equiaxed grains ASTM 7-8



a b c

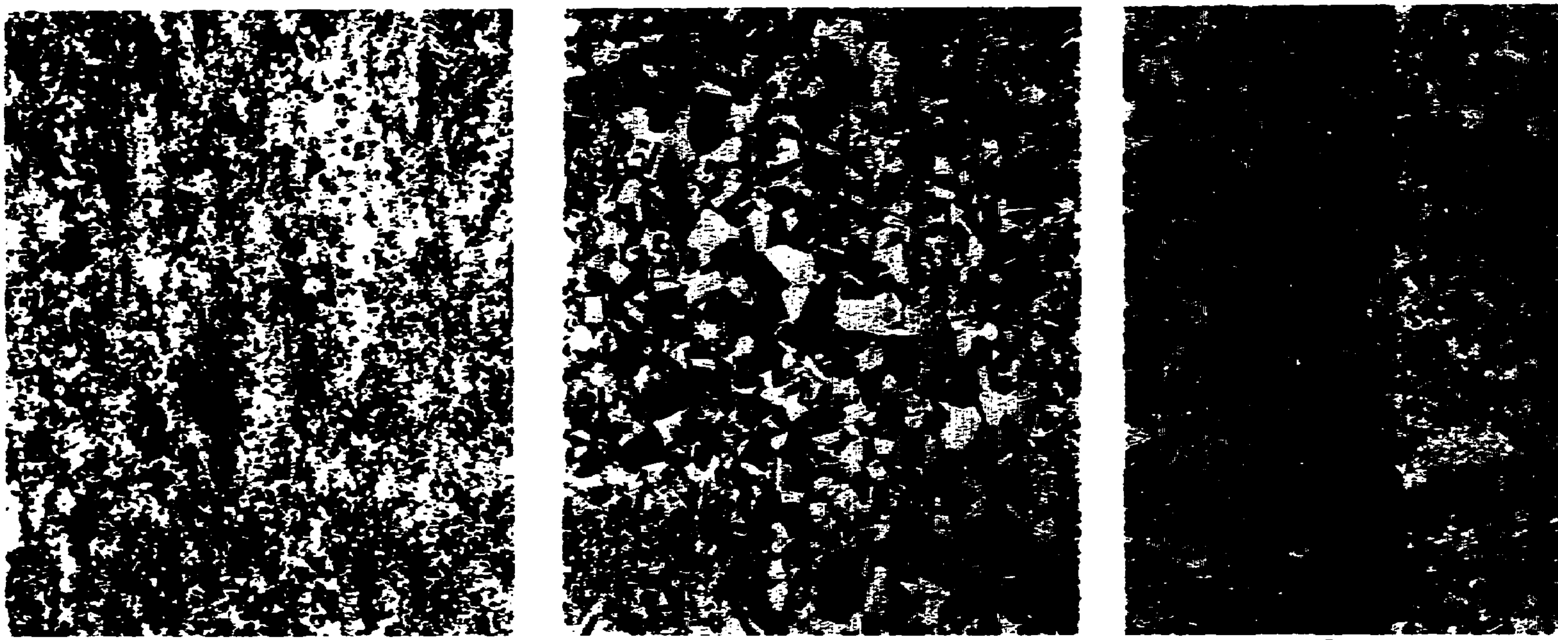
Figure 20 Mag: 100X

Microstructures of forged and heat treated V-5767 (Ti-48Al-1V-.1O_2) gamma alloy bend specimens.

a) 815(1500)/8/AC

b) 1150(2100)/1/Argon Cool +
815(1500)/8/AC

c) 1315(2400)/1/Argon Cool +
815(1500)/8/AC



a b c

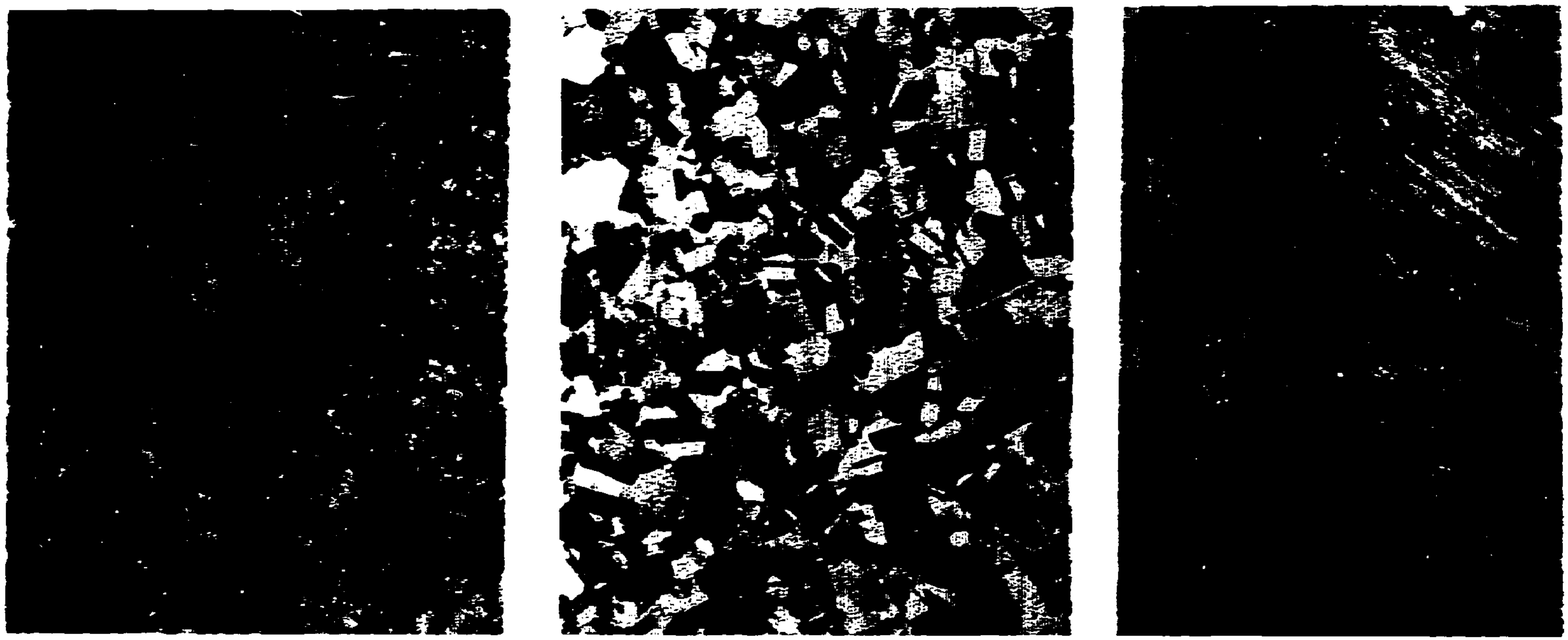
Figure 21 Mag: 100X

Microstructures of forged and heat treated V-5769 ($Ti-48Al-1V-.1C-.1O_2$) gamma alloy bend specimens.

a) 815(1500)/8/AC

b) 1150(2100)/1/Argon Cool +
815(1500)/8/AC

c) 1315(2400)/1/Argon Cool +
815(1500)/8/AC



a b c

Figure 22 Mag: 100X

Microstructures of forged and heat treated V-5770 ($Ti-48Al-.5V-.1O_2$) gamma alloy bend specimens.

a) 815(1500)/8/AC

b) 1150(2100)/1/Argon Cool + 815(1500)/8/AC

c) 1370(2500)/1/Argon Cool + 815(1500)/8/AC

Table 15

Bend Test Results For
Forged and Heat Treated Gamma Alloys

Alloy	Heat Treatment °C(°F)/Hrs.	Test Temp. °C(°F)	0.2% Yield Strength MPa (Ksi)	Max. Fiber Stress at Fracture MPa (Ksi)	Max. % Strain (Ductility)
V-5767 ($Ti-48Al-1V-.1O_2$)	As-forged	RT	620 (89.9)	670 (97.3)	1.05
	815(1500)/8/AC	RT	623 (90.3)	662 (96.0)	1.00
		260 (500)	549 (79.6)	654 (94.8)	1.52
	1150(2100)/1/Argon Cool* +	RT	423 (61.4)	430 (62.4)	1.03
	815(1500)/8/AC		414 (60.0)	533 (77.3)	1.65
	1315(2400)/1/Argon Cool +	RT	524 (76.0)	625 (90.6)	1.28
	815(1500)/8/AC	260 (500)	446 (64.7)	640 (92.8)	1.16
$Ti-48Al-1V-.03O_2$ (MERL Ingot)	As-forged	RT	740 (107.3)	817 (118.5)	1.40
V-5769 ($Ti-48Al-1V-0.1C-0.1O_2$)	As-forged	RT	1017 (147.5)	1086 (157.5)	1.50
	815(1500)/8/AC	RT	1040 (150.8)	1117 (162.0)	1.38
		260 (500)	954 (138.4)	1120 (162.4)	2.12
	1150(2100)/1/Argon Cool +	RT	836 (121.2)	966 (140.1)	1.56
	815(1500)/8/AC	260 (500)	730 (105.9)	867 (125.8)	1.66
	1315(2400)/1/Argon Cool +	RT	644 (93.4)	858 (124.5)	1.98
	815(1500)/8/AC*	260 (500)	523 (75.9)	857 (124.5)	2.69
V-5770 ($Ti-48Al-0.5V-0.1O_2$)	As-forged	RT	577 (83.7)	646 (93.7)	1.15
	815(1500)/8/AC*	RT	594 (86.1)	648 (93.9)	0.97
		260 (500)	530 (76.8)	641 (93.0)	1.56
	1150(2100)/1/Argon Cool +	RT	466 (67.6)	529 (76.7)	1.05
	815(1500)/8/AC	260 (500)	414 (60.1)	507 (73.5)	1.28
	1370(2500)/1/Argon Cool +	RT	478 (69.3)	607 (88.1)	1.43
	815(1500)/8/AC	260 (500)	408 (59.2)	620 (90.0)	2.04

*Selected for tensile and creep testing.

- Increasing the oxygen content to 0.13% (V-5768) resulted in about a 140-175 MPa (20-25 ksi) increase in yield and ultimate strengths with no change in ductility. The low oxygen content MERL ingot showed a measurable increase in ductility, especially reduction in area, at room temperature and up to about 538C (1000F). Yield and ultimate strengths of the low oxygen content ingot were about midway between those of V-5767 and V-5768.

- Direct aging at 815C (1500F) of V-5770 (Ti-48Al-0.1C-.1O_2) resulted in essentially equivalent properties compared to V-5767 (Ti-48Al-1V-.1O_2) which had been solution treated and aged.

- The V-5769 (Ti-48Al-1V-1C-.1O_2) properties were comparable in strength and ductility to those of the high oxygen content V-5768 material even with the higher solution treatment and semi-acicular microstructure. This alloy was judged to have the best overall combination of strength and ductility.

Notched tensile data of heat treated specimens are given in Table 17. For the baseline alloy, the average strength was 209 MPa (29 ksi). Thus, the ratio of notched to smooth strength was 0.62. Both high (V-5768) and low oxygen content compositions exhibited notch strengths higher than baseline but gave calculated notch/smooth strength ratios of 0.72 and 0.80, respectively. The latter value probably reflects the somewhat higher ductility level of the low oxygen material. The baseline V-5767 specimens which had been thermally treated in a hydrogen atmosphere Sievert's apparatus had a notched strength of 304 MPa (44.1 ksi) which gave a notch/smooth ratio of 0.65.

c. Room Temperature Notched Rupture

Room temperature notched rupture testing was conducted using the baseline V-5767 heat before and after attempts at hydrogen charging in a Sievert's apparatus. Chemical analysis showed 28 ppm and 20 ppm for the base and charged specimens, respectively. The specimens were initially loaded at 103 MPa (15 ksi). The load was stepped up 70 MPa (10 ksi) every five hours until rupture occurred. The data (Table 18) showed that the rupture lives and stresses were equivalent for both sets of specimens, and confirmed that the gamma alloy did not absorb any hydrogen.

d. Creep Rupture

Creep rupture data for the three alloy compositions are given in Table 19 and plotted in a Larson Miller format in Figure 23. The carbon containing alloy (V-5769) appears to have nearly a 42C (75F) advantage in life over the V-5767 baseline alloy.

Table 16

Smooth Tensile Test Data For Forged and Heat Treated Gamma Alloys

Heat/Alloy	Heat Treatment °C (°F)	Test Temp. °C(°F)	0.2% Yield Strength MPa	(Ksi)	Ultimate Tensile Strength[1] MPa	(Ksi)	% Elong.	% Red. of Area
V-5767	1150(2100)/1/Argon Cool +	RT	280	(40.6)	321	(48.0)	1.2	2.2
Ti-48Al-1V-0.1O_2	815(1500)/8/AC	260 (500)	239	(34.7)	318	(46.1)	1.4	1.6
		538 (1000)	227	(32.9)	315	(46.7)	1.7	3.6
MERL Ingot	1150(2100)/1/Argon Cool +	RT	380	(55.1)	421	(61.1)	1.1	-
Ti-48Al-1V-0.05O_2	815(1500)/8/AC	RT	335	(48.6)	423	(61.3)	1.3	5.3
		260 (500)	300	(43.5)	441	(63.9)	3.8	3.5
		427 (800)	307	(44.2)	401	(58.1)	1.9	4.1
		538 (1000)	310	(44.9)	443	(64.3)	3.6	3.3
		650 (1200)	293	(42.5)	432	(62.7)	4.2	4.1
V-5768	1150(2100)/1/Argon Cool +	RT	447	(64.8)	502	(72.8)	1.1	2.1
Ti-48Al-1V-0.13O_2	815(1500)/8/AC	260 (500)	382	(55.4)	457	(66.3)	1.5	0.9
		427 (800)	385	(55.8)	454	(65.8)	1.3	0.7
		538 (1000)	365	(52.9)	487	(70.7)	1.8	4.6
		650 (1200)	370	(53.7)	539	(78.2)	4.1	6.1
V-5769	1315(2400)/1/Argon Cool +	RT	356	(51.7)	472	(68.4)	1.5	3.5
Ti-48Al-1V-0.1C-0.1O_2	815(1500)/8/AC	RT	385	(55.8)	507	(73.6)	1.6	2.4
		260 (500)	319	(46.2)	514	(74.8)	2.7	2.9
		427 (800)	325	(47.2)	452	(65.6)	1.9	2.2
		650 (1200)	303	(44.0)	423	(61.2)	7.2	7.9
V-5770	815(1500)/8/AC	RT	343	(49.7)	365	(52.9)	0.9	1.8
(Ti-48Al-0.5V-0.1O_2		RT	350	(50.8)	392	(56.9)	1.1	1.9
		260 (500)	310	(44.9)	310	(44.9)	0.2	0
		427 (800)	321	(46.6)	405	(58.7)	1.7	2.2
		538 (1000)	310	(44.9)	409	(59.3)	2.1	2.1
		650 (1200)	301	(43.7)	486	(70.5)	7.9	7.6

(1) Failure Stress - Specimens broke reaching ultimate strength.

Table 17

Effect of Interstitial Element Content on the Room Temperature Notched (Kt=3.9) Tensile Strength of Ti-48Al-1V Alloy(1)

Heat Identity	Interstitial Content, ppm H_2	O_2	Notched Tensile Strength MPa (Ksi)	Average
V-5767 (Baseline)	28	900	221 (32.1)	206 (29.0)
MERL Ingot Low Oxygen	30	300	338 (49.0)	338 (49.0)
V-5768 High Oxygen	25	1300	319 (46.2) 405 (58.8)	362 (52.5)
V-5767(2) Hydrogen Charged	20	900	304 (44.1)	304 (44.1)

(1) Heat treatment 1150C (2100F)/1/ + 815C (1500F)/8/AC

(2) Baseline heat treatment after thermal H_2 charging

Table 18

Effect of Hydrogen Charging
on the Room Temperature Notched (Kt=3.9)
Rupture Properties of Ti-48Al-1V

Heat Identity	Hydrogen Content, ppm	Time To Rupture, Hrs.	Final Stress, MPa (R) (Ksi)
V-5767	28	15.1	310 (45)
Baseline	28	15.1	310 (45)
V-5767 (1)	20	14.6	241 (35)
H_2 Charged	20	17.5	310 (45)

(1) Exposed in Sievert's apparatus and given baseline heat treatment, 1150C (2100F)/1/Argon Cool + 815C (1500F)/8/AC.

(2) Initial stress 105 MPa (15 ksi) increased 70 MPa (10 ksi) every five hours.

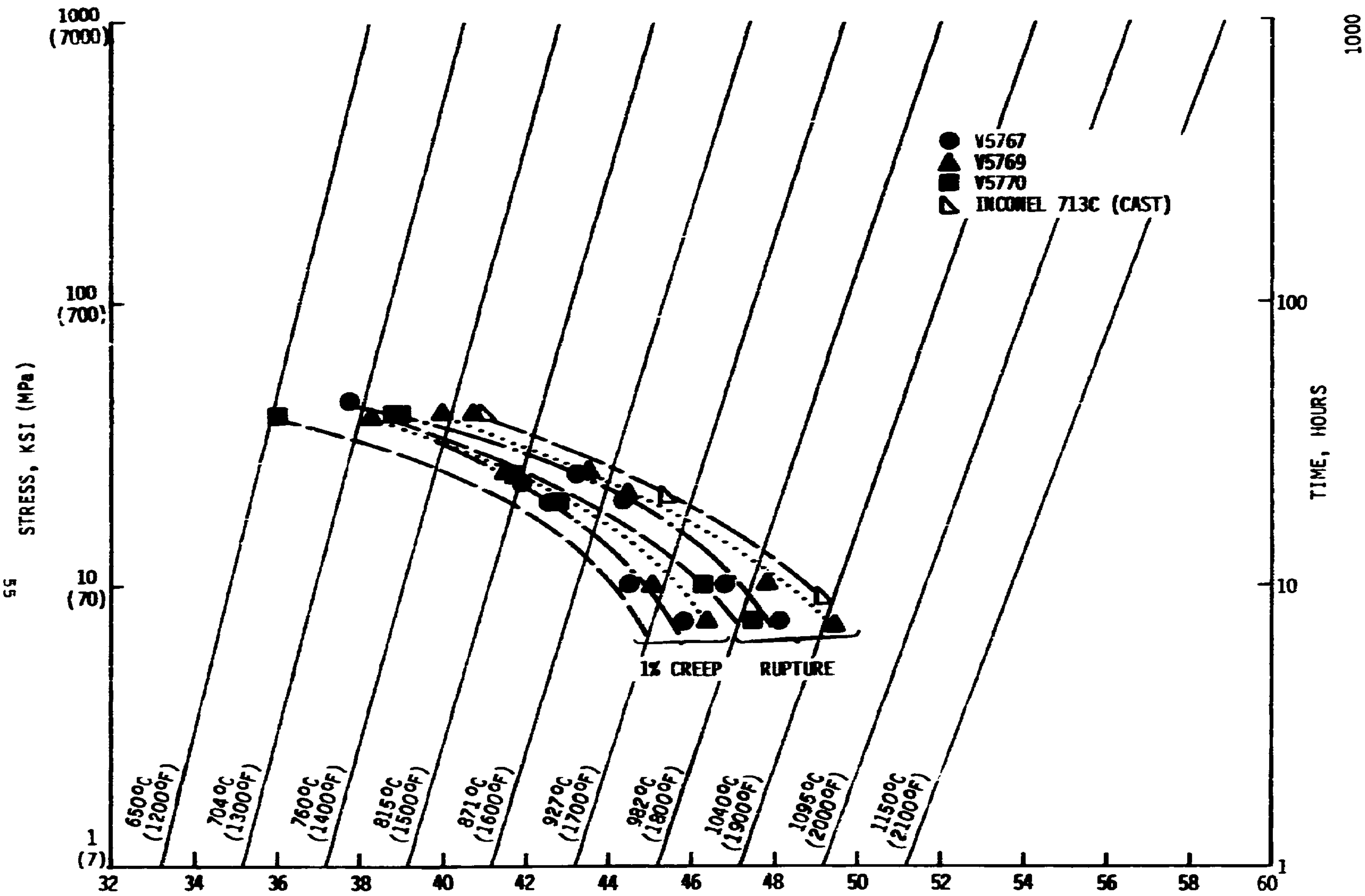
Table 19

Creep Rupture Test Results For
Forged and Heat Treated Gamma Alloys

Alloy	Heat Treatment °C(°F)/Hrs/	Test Conditions °C/MPa (°F/Ksi)	Time, Hours 1%	Time, Hours Rupture	Post Test %EL	Post Test %RA
V-5767 Ti-48Al-1V-.1O_2	1150(2100)/1/Argon Cool + 815(1500)/8/Air Cool	704/345(1300/50)	-	On loading	(1)	0.8
		760/310(1400/45)	0.3	1.9	7.4	7.5
		815/172(1500/25)	20.0	100.5	(1)	10.0
		871/138(1600/20)	6.0	31.0	19.1	31.3
		927/70(1700/10)	5.0	58.0	(1)	(1)
		982/52(1800/7.5)	2.2	18.9	55.6	70.0
V-5769 (Ti-48Al-1V-.1C-.1O_2	1315(2400)/1/Argon Cool + 815(1500)/8/Air Cool	704/276(1300/40)	59.5	581.7 (2)	8.2	10.8
		760/276(1400/40)	4.5	82.4	28.6	53.8
		815/172(1500/25)	15.0	198.1	33.5	57.4
		871/138(1600/20)	3.8	37.1	20.5	18.9
		927/70(1700/10)	7.0	138.7	(1)	(1)
		982/52(1800/7.5)	3.0	72.4	(1)	(1)
V-5770 Ti-48Al-0.5V-0.1O_2	815(1500)/8/Air Cool	704/276(1300/40)	3.0	103.7	49.6	60.0
		760/276(1400/40)	0.3	6.2	44.6	53.3
		815/172(1500/25)	1.5	15.7	33.8	47.5
		871/138(1600/20)	0.5	6.0	20.3	81.8
		927/70(1700/10)	2.1	23.4	97.8	93.9
		982/52(1800/7.5)	0.2	9.1	79.5	(1)

(1) Unable to measure due to nature of fracture.

(2) Uploaded to 310 MPa (45 ksi) at 524.6 hours.



$P = (T+460)(C + \log_{10} t) \quad C = 20$

FIGURE 23

Time to 1% creep elongation and rupture for forged and heat treated gamma alloys compared to Inconel 713C.

The advantage was closer to 55C (100F) when compared to the V-5770 alloy. This probably is a result of the fact that V-5770 (Ti-48Al-0.5V-.1O_2) was direct aged at 815C (1500F) and not solution treated. As discussed in the heat treatment section, direct aging at this temperature results in lower creep life.

For comparison purposes, the creep rupture life of a cast nickel base alloy, Inconel 713C, is plotted on a density corrected basis in Figure 23. It can be seen that the V-5769 specimens approach the life of this cast nickel base alloy when compared on this basis.

It may be perceived that the present stage of development of the gamma alloys is less advanced than the alpha two type systems. Both the processing difficulties and the somewhat low tensile properties point to areas in which improvements are needed before widespread application could be contemplated. Cracking and extensive porosity in some of the ingots was unexpected, although material produced subsequently has not exhibited the latter phenomenon. It was initially considered that the vanadium addition may have been responsible for the porosity in some unknown way but too high a helium partial pressure during melting is a more likely explanation. Although the porosity was virtually eliminated by subsequent isostatic pressing, it is possible that residual defects could have influenced the mechanical properties of the alloys (limited fractographic analyses of failed specimens has failed to confirm the speculation to date).

The tensile properties measured on the various alloys were not as good as anticipated from laboratory heat data. However, we may note that these sections were probably the largest processed and evaluated in the overall program, and scale-up problems should not have been a surprise. Both strength and ductility tended to be low, although the data base with which comparisons could be made was rather restricted. The absolute values of the uniform elongation were probably the best set of data that has been reported to date, but higher values were anticipated at temperatures above 260C (500F). The influence of oxygen was not dramatic. Strength levels increased with oxygen and ductility, especially the reduction in area values, tended to be small at low temperatures. However, oxygen had virtually no effect on the notch tensile strength ratios of $\sim$0.8 being measured for both the low and high oxygen heats. The influence of heat treatment on properties is also not clear from this study as a chemistry variation (carbon addition) was included together with heat treatment at higher temperatures. The properties of this alloy solution treated in the two

phase ($\beta + \gamma$) region appear generally attractive. As the nature of the transformation in these alloys has become clearer, an option that could be useful would be to add more beta stabilizer (V, Mo, etc.) to reduce the transformation temperature to ∿1150C (2100F).

Creep rupture properties of the alloy series were relatively constant although the carbon containing alloy displayed improved capability. Equivalence with IN713C (on a density corrected basis) is all that is required for the initial component program for these alloys, the construction and test of JT9D low pressure turbine blades. Earlier work on contract F33615-75-C-1166 has shown that should further improvements be needed, both higher solution temperatures and/or the addition of tungsten can be used.

SECTION V

SUMMARY AND CONCLUSIONS

1. Based on the test data obtained in the current contract, the major findings may be stated as follows.

2. Alpha-Two Alloys

- Beta forging of large ingots of these alloys using conventional titanium alloy practice was demonstrated.

- Of the three alloys studied, no one composition showed any significant advantage when heat treated to a fine Widmanstatten microstructure. Ternary Ti-Al-Nb compositions were slightly better than the Ti-Al-Nb-V alloy, but not enough to eliminate the latter from consideration as a useful engineering material.

- Interstitial effects were found to parallel behavior of conventional titanium alloys; increased hydrogen levels reduced notched rupture strength and higher oxygen content lowered ductility and increased strength at low and moderate temperatures. When oxygen content was raised from 0.03 to 0.13%, the beta transus temperature was increased 55C (100F).

- Room temperature tensile ductility of Ti-25Al-14Nb specimens containing <500 ppm oxygen was more than twice that of similarly processed Ti-25Al-14Nb specimens with 800-1000 ppm of oxygen.

- The original goals established in 1975 were to develop a material equivalent in tensile properties to beta processed Ti-6Al-2Sn-4Zr-2Mo with creep properties equivalent to Inconel 713C. The first goal was met, but the current alloys fall short of the second. However, the property levels represent very significant improvements over the original alloys studied.

3. Gamma Alloys

- The state of development of the gamma intermetallic compounds is less advanced than for alpha-two alloys as reflected in the processing difficulties encountered.

- A systematic study of the heat treatment/microstructure relationship was performed and the grain growth behavior of forged material explained.

- The tensile properties were lower than anticipated based on previous data from laboratory heats. It is possible that the unhealed ingot defects contributed to the low ductility values.

- The measured effect of interstitial elements was not great in the gamma alloys. Hydrogen could not be charged into the material and increased oxygen content caused only a slight increase in strength and decrease in ductility.

- Creep-rupture properties were enhanced by small carbon additions and/or by using a high solution temperature resulting in some acicular transformed structure. This alloy was essentially equivalent to Inconel 713C in creep resistance.

- Properties of the Ti-48Al-1V alloys appear to be adequate for production and testing of JT9D turbine blades in a current ManTech program.

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